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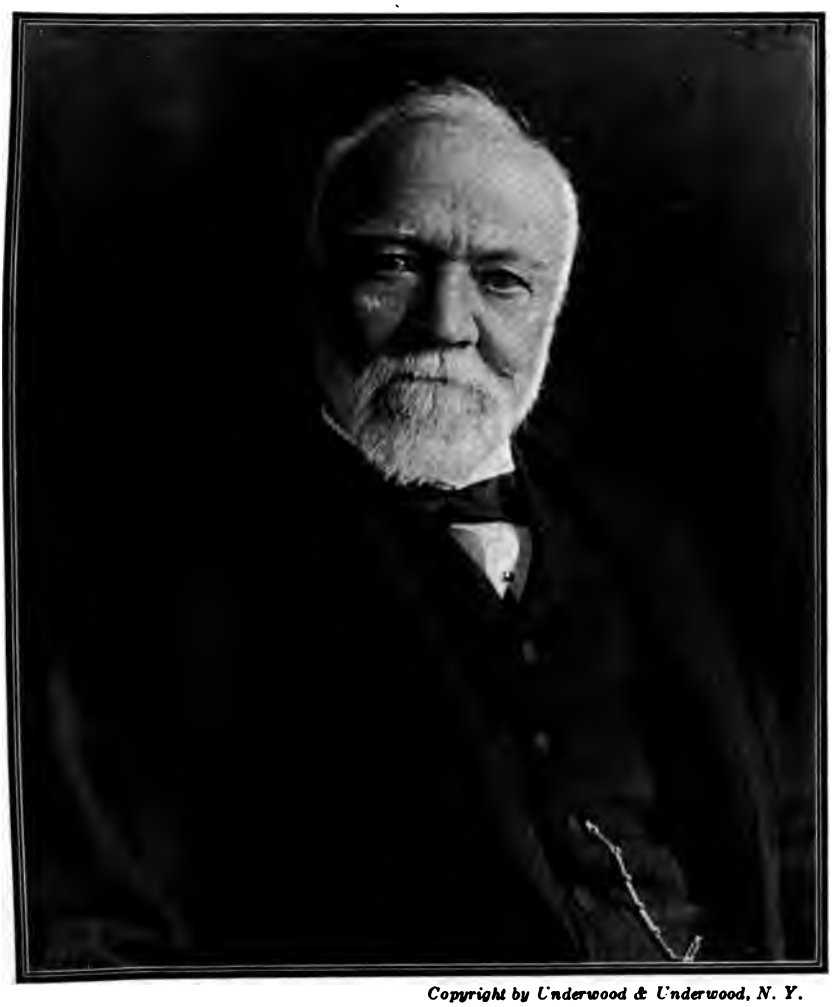

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ANDREW CARNEGIE

TRANSACTIONS

OF THE

AMERICAN INSTITUTE OF MINING AND METALLURGICAL ENGINEERS

(INCORPORATED)

VOL. LXII

CONTAINING PAPERS AND DISCUSSIONS ON IRON AND STEEL, PRESENTED AT THE COLORADO MEETING, SEPTEMBER, 1918, AT THE NEW YORK MEETING, FEBRUARY, 1919, AND AT THE CHICAGO MEETING, SEPTEMBER, 1919.

NEW YORK, N. Y.
PUBLISHED BY THE INSTITUTE
AT THE OFFICE OF THE SECRETARY
29 WEST 39TH STREET
1920

524259

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THE MAPLE PRESS YORK PA

PREFACE

In this volume are the papers and discussions on Iron and Steel that were presented at the Colorado meeting in September, 1918, at the New York meeting in February, 1919, and at the Chicago meeting, September, 1919. An account of the Proceedings of the Chicago meeting is also included:

With the publication of this volume, *Bulletins* 140 to 146 are superseded by the *Transactions*.

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PROCEEDINGS OF THE ONE HUNDRED AND TWENTIETH MEETING AT CHICAGO

The one hundred and twentieth meeting of the Institute was held at Chicago, Sept. 22 to 26, 1919, inclusive, and was in every way a success, although the steel strike against the United States Steel Corpn. prevented many engineers from attending the steel session and made necessary the annulment of the trip to the steel works at Gary on Tuesday.

All of the meetings were held at the Congress Hotel, with the exception of the session on Non-ferrous Metallurgy which was held at the Coliseum where the Fifth National Exposition of Chemical Industries was in progress. This session was a joint one with the American Electrochemical Society.

The total attendance, as shown by the registration, was 850.

The session on Mine Taxation on Monday morning developed so much discussion that it ran into two sessions on Tuesday. This important subject created such widespread interest among both the members present and the members who were absent, as developed by letters and telegrams, that President Winchell appointed a committee of twenty-four to go to Washington, at the invitation of the Commissioner of Internal Revenue, for a two-day session beginning Oct. 6.

The symposium on Sulfur in Coal and the one on Pyrometry were most successful; the papers and discussions of the latter have been published in a special volume. All of the sessions were well attended and the papers, which were of unusual merit, brought out valuable discussions.

THE SMOKER

The smoker, Monday evening, was an unusual success. It was held in the beautiful gold room of the Congress hotel, which was very elaborately decorated with American flags. More than 500 were present. The gallery was filled with ladies, who responded to the request of the song leader and helped in many ways to grace the occasion. The entertainment was a varied one and was original with the Chicago Committee, which deserves great credit for the unusual success of the event. Although it was mostly funmaking, Dr. F. B. Cottrell and George S. Rice of the Bureau of Mines gave some very interesting talks, illustrated by flags and moving pictures, of their trip through Europe since the armistice. Mr. Rice's talk dealt particularly with the coal and iron regions and Doctor Cottrell's was devoted almost solely to that part of his trip in Northern Italy where he was investigating the application of volcanic steam to power purposes. This steam is derived from boracic-acid springs and helium is secured as a byproduct.

FOREIGN GUESTS AT THE INSTITUTE MEETING

There were two official representatives of Foreign Governments as guests at the Chicago Meeting.

One was Fredrik Goransson, managing director of the Sandvikens Steel Works, who represented the Jernkonteret, the Iron and Steel Institute of Sweden, and who came to this country especially for our meeting.

The other distinguished guest was Commander Toublet, who represented the French Embassy. Commander Toublet delivered a most polished address at the Banquet, dealing with the felicitous relations existing between this country and France for the last 150 years. Commander Toublet was accompanied by Mrs. Toublet, who won the hearts of all the ladies present, whose guest she was.

THE BANQUET

The banquet Wednesday evening was attended by about 700. The list of speakers was unusually brilliant and each one proved to be so interesting that it was nearly midnight before the dancing began. Capt. Robert W. Hunt was toastmaster. Captain Hunt is about to celebrate his eighty-first birthday but is blest with unusual vigor and it was a great pleasure to all his friends to honor him at this time. His selection as toastmaster was a happy one. The guest of the evening, if among so many stars one could be selected from the rest, was Charles M. Schwab, who said nothing but the urgent telegram from his dear old friend, Captain Hunt, could have drawn him away from his seclusion on his Pennsylvania farm to attend the meeting in Chicago. Mr. Schwab's address was, as usual, scintillating with wit and humor; and while he did not directly or indirectly allude to the steel strike he did at the end of his talk discuss serious things and gave it as his opinion that any effort to legislate or control the high cost of living would be futile until this country got down to a basis of an honest day's work for an honest day's wage.

Address of Horace V. Winchell

On behalf of the members and directors of the American Institute of Mining and Metallurgical Engineers, I desire to express our appreciation of the diligent labors and many thoughtful courtesies of the Chicago members which have culminated in this splendid gathering here tonight. We have enjoyed the cordial atmosphere of welcome by which we have been surrounded, we have profited by the many interesting discussions of the past three days, and we shall long remember with gratitude the manifestations of friendship which have attended our brief sojourn in Chicago.

We look upon these, our annual mid-year gatherings in different sections of the country, as one of the most effective agencies in the forming of acquaintance with many whom we might otherwise never meet. In this way are brought together men of common aims and similar lines of investigation, and through the interchange of ideas and the friction of personalities are often developed a warmth of interest and a fruition of conception which lead not only to lifelong friendships but to results of importance in our social and economic development. Those of us who were here in 1893 and listened to the discussions of Posepny's paper on the genesis of ores received an inspiration and a stimulus which perhaps have played no small part in the vastly greater study and the clearer understanding of that subject which in the past quarter of a century have been marked by a progress greater than that of all antecedent time. Nor is this the only epoch-making monograph which has first been promulgated at our meetings. I have no doubt whatever that some of the ideas advanced in our technical sessions this week contain germs of

thought susceptible of great future elaboration and that these germs if not already fully developed will bear rare fruit in succeeding years.

May we not therefore congratulate ourselves and thank our hosts in this great metropolis because of the record-breaking attendance here this evening? Truly a high mark has been set in the history of Institute events.

In glancing backward over the past 25 years we become conscious of many changes, not alone in the scale of production nor in our methods of operation but also in the more intangible but none the less important conception of the relation of the engineer to society in general and more especially to his own particular community. Although it is to be feared that there is some foundation for the statement recently made to the Chairman of the National Service Committee of the Engineering Council by "an acknowledged statesman and keen observer of events" that "Engineers are the most unresponsive citizens that we have * * * * Their aloofness and indifference in all matters outside of their own professional sphere are among the unexplained things in our political life," there are signs of an awakening, and the future is not without promise.[1]

Our Institute itself, an organization which was for many years forbidden by its organic law from exerting any influence on law-making bodies, or even presenting its views as an organized unit of professional men regarding any movement which affected the general community, has recently removed these restrictions and is now able legally and with perfect propriety to appoint its committees and representatives to lend professional advice and assistance to any and all public and governmental agencies which may in any way shape and control legislation and national activities.

Recent events have shown us that the fabric of human relationship is each year more closely interwoven. The strands are ever drawn closer and more firmly and each separate thread is a sensitized nerve alive with electricity and communicating instant warning of danger and news of damage to every other vital part. The development and perfecting of means of communication and transportation and the arrival of an industrial age have made all the world neighbors and just as a fire in your neighbor's house is always a matter of concern to you, so an incipient conflagration of social or economic unrest or disorder in any part of the civilized world has now come to have its effect upon every other part. I have said before, and it cannot be too often repeated, that from the historic moment when Admiral Dewey sailed into Manila Bay and upset the coffee cups of the Spanish Navy we the United States of America are by force of natural law, and whether we wish it or not, partners in the commonwealth of nations. We may legislate against it; we may make treaties with reservations, or no treaties at all; we may set up barriers and lock the doors against all foreigners, black, white or yellow; but if because of our acts or for any other reason, these excluded and distant peoples engage in conflict among themselves, we shall again, in spite of our wishes, our barriers and our padlocks, be drawn into the fray.

Nor is it in time of war alone that we find ourselves indissolubly harnessed up with the rest of mankind. Each smallest fluctuation in the annual production of the farms of Europe or of Argentine is reflected

[1] *Eng. and Min. Jnl.* (Sept. 13, 1919) 439.

in our markets; each drouth or famine, each outburst of Bolshevism, each overthrow of a tyrannical government in any part of the globe affects each and every one of us, down to the youngest babe in its cradle or the poorest paid worker in a sweatshop. Social diseases of the body politic are like the influenza, they are pandemic, they are in the air and are not excluded by bolts nor barred gateways, by statutes nor standing armies.

Nor may we blind our eyes to the fact that at the present time, under the order of divine providence, the United States is the only great nation whose resources are not only unimpaired but incomparably greater than ever before; that the nations of Europe are exhausted and almost bankrupt; that it is our privilege and our duty to them as well as to ourselves to go to their rescue. This was made so clear by our illustrious member, Herbert Hoover, in his address at New York last week, that it requires no emphasis from me. I presume each of you will soon receive that address in printed form.[2] I advise you to learn it by heart and loan it to all of your friends. I wish that it could be read by every intelligent person between the Atlantic and Pacific; that it could be used as a text by every minister and as a reading lesson for a week in every high school. It emanated from a heart wrung by the woes of suffering millions and from a brain clear enough and broad enough to analyze the causes for Europe's condition and to perceive clearly the only remedy. It is a source of inspiration to us and one of supreme gratification to all engineers that in a time of crisis, in a period that tries men's souls, at a moment when the real man comes to the front and assumes that position of leadership to which, by his mental attributes and moral courage he is naturally entitled, it should have been a mining engineer to whom the world turned for relief, and that for his trusted and most effective coadjutors, he chose many other engineers.

Does this not suggest to us that engineers are needed in the settlement of the world's affairs? Do we not realize that there is prevalent as an aftermath of war, and the artificial conditions created by it, a wide-spread feeling of discontent? Are we not still, and more than ever, ready to apply to the solutions of post-war problems, the mental poise and balanced judgment which are a part of the equipment of the trained engineer? Shall we be so modest and retiring or so engrossed in our technical duties as to forget or neglect our duties as citizens?

Just at this point my subliminal ear detects a murmur. What do you say? "We cannot all be Hoovers or Presidents or Members of Congress?" No, of course not! But you can help to make or unmake them, to influence them, and to shape the policies of government and the destinies of nations. Why is President Wilson touring the country today? Is it not to shape public opinion? He knows that the world is ruled by ideas, and that there is no greater force than public opinion. Suppose all the engineers of this broad republic should unite and devote all of their spare time and energy for six months in advocating some change in our statutes which clearly appealed to their sense of equity as promoting the public welfare, do you not suppose they could overcome the apathy and selfish inertia of our chief law-making body? Of course, they could. Well, it would not be you or I, but all of us, who would have done it. It would be the result of coöperation and organized effort, in which each individual pulled in the same direction and not at cross-purposes with the rest.

[2] *Bull.* 154 (October, 1919) xi.

Now, this is all trite; and in one form or another has been said many times. You want perhaps to know just what particular move to make next, and just how and where to begin. And these questions I cannot answer; but I may, with your permission, make a few suggestions.

There are problems constantly before us of public character and yet of direct and vital importance to each one of us engineers. Some of these problems are comparatively local in their sphere of bearing and application, some are nation wide, and some of them are international in scope. We have, for example, in each state and county, local taxation questions. As applied to mines and quarries, to mills and smelters, and to many other industrial plants, the valuation of properties for taxation involves questions which require engineering ability of high order, and keen analysis by the best brains of the country. The same problems are now presented in connection with the Federal Income Tax law. Indeed, there are few subjects of wider general bearing and importance, nor any which make more pressing demand upon the engineering talent of the land. That we have not been consulted in the framing of the law is perhaps largely due to our lethargy and apparent indifference. We are now confronted with its interpretation and proper application, and should not fail to give to the governmental authorities the benefit of our most serious consideration and competent advice. We are glad to be called into conference; and must willingly embrace the opportunity to aid in placing all interests involved on the most nearly equitable basis.

Then there is the ever-present and just now unusually inflamed antagonism between the employee and the employer. I know of no more promising work for engineers with their peculiar adaptation for studying the elements of a complicated situation, for determining the relative importance of its factors, for picking out the particular operating defects, and for suggesting solutions to difficulties. Conflicts generally arise through lack of mutual understanding and consequent lack of confidence. Where there is perfect understanding among fair-minded men there will be confidence; and where there is confidence, difficulties can be adjusted without strife. It seems to me that engineers may be of inestimable service to all classes by exerting the influence, which they undoubtedly possess, toward the education of each of the hostile elements as to the real necessities and actuating motives of the other. We are consumers; we are a portion of the public which always suffers in these industrial conflicts. We are coming to believe that no element in the community should be permitted to throw our entire industrial machine out of operation: that some control should and must be exercised to prevent these recurring disasters to the peace and prosperity of the country. The subject is one which demands the attention of all thoughtful and intelligent citizens and is to be settled, if at all, only by concerted effort.

In broader fields we see the need of engineering brains and attention in the worldwide control and distribution of mineral resources. Recent thoughtful discussions by Messrs. Leith and Spurr have shown us that the United States has important duties to its own people and to the people of other countries in connection with the commercial control of minerals. It will be generally admitted that, until very recently, our people have been so busy in developing our own resources that we have paid little attention to those of other countries. It is also a fact that our policy of insularity has tended to discourage investments abroad. The studies of world conditions which we have been forced to make within the past

few years have made it clearly evident that we must wake up and protect ourselves and our industries in their relation to the raw materials for manufacture and worldwide distribution. Our resources must not be dominated by foreign capital and we must acquire our necessary share of those which can be controlled by us abroad. As Spurr wisely says:[3]

"Such mineral wealth as we possess in an exportable surplus must be managed for our best advantage. Such minerals as we do not possess in quantities sufficient for our own needs must be secured to us so far as possible by a definite and intelligent governmental policy."

Other governments have already passed laws debarring us from owning or operating oil-producing properties and mines of various sorts within their territories. We have done little or nothing along these important lines of security and protection for the future.

Finally, it is for us to remember that we are not only engineers but Americans; that "America is worth saving not only as a land in which men and women may be free and increasingly prosperous, but as a land and a government under which character can be built, individual capacity given opportunity for free exercise, and coöperation on the widest scale promoted not only for private advantage but for the public good."[4]

Let us not forget that "socialism is the twin brother of autocracy and, like autocracy, it is the deadly enemy of republicanism and of individual liberty." Let us stand firmly for the principles of equal rights and justice for all and have full confidence in the destiny of our country. Thus alone shall we live up to the full measure of our profession and our opportunities.

ALL-DAY EXCURSIONS

The only technical session on Thursday was a continuation of the Symposium on Pyrometry. The members not attending the symposium made the trip to La Salle. They were welcomed by the Mayor, to whose speech Major Arthur S. Dwight briefly responded. After luncheon the party divided and was taken by automobile to the various chosen points of interest, such as the cement works, the various zinc smelters, the coal mines, and underground cement-rock mines, where underground steam shovels were operating, etc. A special scenic trip was arranged for the ladies. The party returned by special train to Chicago, arriving at 7:30.

Those desiring to visit the coal mines at Franklin and Macoupin Counties left by train that evening on a special car for Benton, and were piloted through the field by Mr. Putnam, superintendent of the Old Ben Coal Corpn., Mr. Carroll, assistant general superintendent of the Chicago, Wilmington & Franklin Coal Co., and Carl Scholz, general manager of the Vallier Coal Co. Through the courtesy of the various railroads the visitors were furnished a special coach which took them to the various mines with the least delay and enabled them to see more than would be otherwise possible. Mines 8 and 9 of the Old Ben Coal Corpn., at West Frankfort, were first visited; thence the party proceeded to Orient mine of the Chicago, Wilmington & Franklin Coal Co., where luncheon was

[3] *The Scientific Monthly* (July, 1919) 80.

[4] Nicholas Murray Butler. Address before Cincinnati Commercial Club, Apr. 19, 1919.

served, and then proceeded to Vallier, where both the top and the underground workings were inspected. The party returned to Benton and from there proceeded in automobiles to inspect the washery plant of the United States Fuel Co. at that point; and boarded the train reaching Chicago Saturday morning. Several left Benton for their homes without returning to Chicago.

COAL EXCURSION PARTY AT ORIENT MINE.

On Friday 110 people took the trip to North Chicago and Milwaukee. The train stopped at the Fansteel plant at North Chicago. Arriving at Milwaukee about 12:30, the party was entertained at luncheon by the Allis-Chalmers Co. in its club house. Afterwards the plant was visited. Two of the three ladies who accompanied the excursion were furnished a car by the Allis-Chalmers Co. to see the various points of interest in the city; the other lady preferred to remain with her husband and see the Allis-Chalmers plant. On the floor of the plant was mining machinery of all kinds that was to go to all parts of the world.

CHAIRMEN OF COMMITTEES

It was a very strenuous week and a very successful meeting. The chairmen of the committees to whose labors this success was due are as follows: Ladies, Mrs. Chas. H. MacDowell; Finance, Carl Scholz; Publicity, F. G. Fabian; Papers, H. H. Stoek; Excursions, L. V. Rice; La Salle District, J. A. Ede; Industries, G. M. Davidson; Guides, F. W. DeWolf; Reception, Robert W. Hunt; Hotel, F. T. Snyder; Smoker Entertainment, C. W. Gennet, Jr.; Banquet, F. G. Fabian.

TECHNICAL SESSIONS

Mine Taxation

On Monday morning, Sept. 22, a session on Mine Taxation was held in coöperation with the Internal Revenue Dept., U. S. Treasury, Mr. R. C. Allen presiding. This was followed by two adjourned sessions, at which Mr. R. C. Allen and Mr. A. D. Brokaw presided. The following paper was presented:

Federal Taxation of Mines. By L. C. Graton. (Presented by the author; discussed by R. V. Norris, W. O. Hotchkiss, W. P. Beldon, Paul Armitage, R. C. Allen, E. F. Brown, William Kelly, A. D. Brokaw, F. W. Sperr, C. H. Benedict, E. P. Griffitts, W. E. Forsythe, and the author).

Coal and Gas

The first session on Coal and Gas was held on Monday morning, Sept. 22; the second, on the afternoon of the same day. Mr. Carl Scholz presided at both sessions. The following papers were presented:

Research in the Coal-mining Industry. By E. A. Holbrook. (Presented by Mr. Reed.)

Some Factors that Affect the Washability of a Coal. By Thomas Fraser and H. F. Yancey. (Presented by Mr. Fraser.)

A Use Classification of Coal. By George H. Ashley. (Presented by title.)

Distribution of Anthracite. By A. S. Learoyd. (Presented by title.)

Height of the Gas Cap in the Safety Lamp. By C. M. Young. (Presented by the author.)

Engineering Features of Modern Large Coal Mines in Illinois and Indiana. By C. A. Herbert and C. M. Young. (Presented by Mr. Young.)

Gas-producer Practice. By G. S. Brooks and C. C. Nitchie. (Presented by Mr. Nitchie.)

Testing of Coals for Byproduct Coking and Gas Manufacture. By Horace C. Porter. (Presented by the author.)

Coals of Ohio and their Limitations for Byproduct Coke. By Wilbur Stout. (Presented by title.)

Outdoor Substations in Connection with Coal-mining Installations. By H. W. Young. (Presented by the author.)

Non-ferrous Metallography

The session on Non-ferrous Metallography was held at 11 A.M. on Monday, Sept. 22, Mr. G. M. Fritch presiding. The papers at this meeting were all presented by title only to afford opportunity for discussion. The same papers were presented at the meeting of the Institute of Metals Division in Philadelphia, Sept. 29–Oct. 2, and the list of titles will be found in the account of the Proceedings of that meeting, in the next Non-ferrous Volume.

Geology

On Monday afternoon, Sept. 22, a session on Geology was held, Prof. A. N. Winchell presiding. The following papers were presented:

Chrome-ore Deposits in Cuba. By Ernest F. Burchard. (Presented by the author.)

Recent Studies of Domestic Chromite Deposits. By J. S. Diller. (Presented by title.)

Manganese-ore Deposits in Cuba. By Ernest F. Burchard. (Presented by the author.)

Correlation of the Formations of the Huronian Group in Michigan. By R. C. Allen. (Presented by the author.)

Mud Volcanoes in Colombia. By Stanley C. Herold. (Presented by title.)

Magnesite; its Geology, Products and their Uses. By C. D. Dolman. (Presented by title.)

Titaniferous Iron Sands of New Zealand. By V. W. Aubel. (Presented by title.)

Recent Studies of Domestic Manganese Deposits. By E. C. Harder and D. F. Hewett. (Presented by title.)

Milling and Industrial Organization

A session on Milling and Industrial Organization was held on Monday, Sept. 22, at 2 P.M., Mr. F. K. Copeland presiding. The following papers were presented:

Chilean-mill Practice at the Portland Mill. By Luther W. Lennox. (Presented by G. M. Taylor.)

Graphic Metallurgical Control. By H. M. Merry. (Presented by title).

Mill Operations at the United Eastern during 1917–1918. By Wheeler O. North. (Presented by title.)

Crushing Practice at New Cornelia Copper Co. By W. L. Du Moulin. (Presented by title.)

Method of Curtailing Forces at the Copper Queen. By Charles F. Willis. (Presented by T. T. Read.)

Educational Methods at the Copper Queen. By Charles F. Willis. (Presented by T. T. Read.)

Physical Examination Previous to Employment. By Charles F. Willis. (Presented by T. T. Read.)

Iron and Steel

There were three sessions on Iron and Steel. On Tuesday afternoon Prof. J. W. Richards presided; on Wednesday morning, Dr. J. A. Mathews; and at the session on Tuesday evening, Mr. Horace V. Winchell was the presiding officer. The following papers were presented:

Blast-furnace Refractories. By Raymond M. Howe. (Presented by the author.)

Effervescing Steel. By Henry D. Hibbard. (Presented by title.)

Aircraft Steels. By Albert Sauveur. (Presented by title.)

Determining Gases in Steel and the Deoxidation of Steel. By J. R. Cain. (Presented by the author.)

Effect of Time and Low Temperature on Physical Properties of Medium-carbon Steel. By G. A. Reinhardt and H. L. Cutler. (Presented by W. N. Crafts.)

Erosion Tests of Rifle Barrels. By A. E. Bellis. (Presented by title.)

Metallography of Rifle-barrel Steel. By G. F. Butterworth. (Presented by title.)

Manufacture of Steel Rails. By Robert W. Hunt. (Presented by the author.)

Cooling Properties of Technical Quenching Liquids. By N. B. Pilling and T. D. Lynch. (Presented by Mr. Pilling.)

Differential Crystallization in Cast-steel Runner. By Francis B. Foley. (Presented by title.)

Manufacture and Properties of Light-wall Structural Tubing. By H. J. French. (Presented by the author.)

Oxygen in Cast Iron and its Application. By Wilford L. Stork. (Presented by title.)

Graphitization of White Cast Iron upon Annealing. By Paul D. Merica and L. J. Gurevich. (Presented by title.)

Experimental Data Obtained on Charpy Impact Machine. By F. C. Langenberg. (Presented by title.)

Heat Treatment of Cast Steel. By John H. Hall, Arvid E. Nissen and Knox Taylor. (Presented by Mr. Hall.)

Deep Etching of Rails and Forgings.* By F. M. Waring and K. E. Hofammann. (Presented by title.)

Oil

A session on Oil was held on Tuesday evening, Sept. 23, Mr. C. W. Washburne presiding. The following papers were presented:

Irvine Oil District, Kentucky. By Stuart St. Clair. (Presented by title.)

Petroliferous Provinces. By E. G. Woodruff. (Presented by title.)

Investigations Concerning Oil-water Emulsion. By A. W. McCoy, H. R. Shidel and E. A. Trager. (Presented by Mr. Trager.)

Essential Factors in Valuation of Oil Properties. By Carl H. Beal. (Presented by title.)

Application of Law of Equal Expectations to Oil Production in California. By Carl H. Beal and E. D. Nolan. (Presented by A. D. Brokaw.)

Value of American Oil-shales. By Charles Baskerville. (Presented by A. W. Ambrose.)

Sulfur in Coal

Two sessions were held on Sulfur in Coal, on Wednesday morning and Wednesday afternoon, Sept. 24. Prof. H. H. Stoek presided at both sessions. Papers were presented as follows:

Geographic Distribution of Sulfur in the West Virginia Coal Beds. By I. C. White. (Presented by J. W. Paul.)

Occurrence and Origin of Finely Disseminated Sulfur Compounds in Coal. By Reinhardt Thiessen. Illustrated by lantern slides. (Presented by the author.)

Mechanical Separation of Sulfur Minerals from Coal. By J. R. Campbell. (Presented by title.)

Sulfur in Coal—Geological Aspects. By George H. Ashley. (Presented by title.)

Forms in Which Sulfur Occurs in Coal. By A. R. Powell and S. W. Parr. (Presented by Mr. Powell.)

Effect of Sulfur in Coal used in Ceramic Industries. By C. W. Parmelee. (Presented by title.)

Removal of Sulfur from Illuminating Gas. By W. W. Odell and W. A. Dunkley. (Presented by Prof. Hood.)

Low-sulfur Coal in Pennsylvania. By H. M. and T. M. Chance. (Presented by title.)

Low-sulfur Coals of Kentucky. By Willard R. Jillson. (Presented by title.)

Low-sulfur Coal in Illinois. By Gilbert H. Cady. (Presented by title.)

Sulfur in the Coking Process. By S. W. Parr. (Presented by Mr. Powell.)

Commercial Recovery of Pyrite from Coal. By S. H. Davis. (Presented by title.)

Sulfur in Producer Gas. By F. Crabtree and A. R. Powell. (Presented by Mr. Powell.)

Mining and Local Resources

A session on Mining and Local Resources was held on Wednesday morning, Sept. 24. Mr. F. W. DeWolf presided. The following papers were presented:

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* Presented at the June, 1919, meeting of the American Society for Testing Materials, and read here only to afford opportunity for discussion.

Wisconsin Zinc District. By W. F. Boericke and T. H. Garnett. (Presented by Mr. Boericke.)

Mineral Resources of the La Salle District. By J. A. Ede. (Presented by the author.)

New Angles to the Apex Law. By John A. Shelton. (Presented by the author.)

Mining Methods of Alaska Gastineau Mining Co. By G. T. Jackson. (Presented by title.)

Tunnel Driving at Copper Mountain, B. C. (Columbia Section Paper). By Oscar Lachmund. (Presented by title.)

Geology and Mining Methods at Pilares Mine. By W. Rogers Wade and Alfred Wandtke. (Presented by title.)

Wedging Diamond-drill Holes. By O. Hall and V. P. Row. (Presented by title.)

Non-ferrous Metallurgy

On Wednesday afternoon, Sept. 24, a joint session with the American Electrochemical Society was held, on the subject of Non-ferrous Metallurgy. Mr. E. P. Mathewson presided. Papers were presented as follows:

Electric-resistance Furnace of Large Capacity for Zinc Ores. By Charles H. Fulton. (Presented by the author.) Illustrated by lantern slides.

Water and Chlorides in Cement Copper Briquettes. By Edward Keller. (Presented by title.)

Chemical and Electrochemical Problems Involved in New Cornelia Copper Company's Leaching Process. By Henry S. MacKay. (Presented by title.)

Electrolytic Zinc. By C. A. Hansen. (Presented by the author.)

Treating Antimony Ores. By George P. Hulst. (Presented by the author.)

*Pyrometry**

A session on Pyrometry with Special Reference to Iron and Steel Metallurgy was held on Wednesday afternoon, Sept. 24, at which Dr. G. K. Burgess presided and the following papers were presented:

Report of Pyrometer Committee of National Research Council. By George K. Burgess. (Presented by the author.)

Pyrometry in Blast-furnace Work. By P. H. Royster and T. L. Joseph. (Presented by P. H. Royster.)

Pyrometry and Steel Manufacture. By A. H. Miller. (Presented by the author.)

Electric, Open-hearth and Bessemer Steel Temperatures. By F. E. Bash. (Presented by the author.)

Some Thermal Relations in the Treatment of Steel. By Charles F. Brush. (Presented by the author.) Illustrated by lantern slides.

Pyrometry in the Tool Manufacturing Industry. By J. V. Emmons. (Presented by the author.)

Forging Temperatures and Rate of Heating and Cooling of Large Ingots. By F. E. Bash. (Presented by the author.)

The second session on Pyrometry was held on Thursday, Sept. 25, beginning at 10 A.M. Doctor Burgess presided and the following papers were presented:

Temperature. By J. S. Ames. (Presented by title.)

Standard Scale of Temperature. By C. W. Waidner, E. F. Mueller and Paul D. Foote. (Presented by Mr. Mueller.)

Metals for Pyrometer Standardization. By C. W. Waidner and George K. Burgess. (Presented by title.)

Fundamentals of Pyrometry. By C. E. Mendenhall. (Presented by the author.)

*In coöperation with the National Research Council and the National Bureau of Standards.

Thermoelectric Pyrometry. By Paul D. Foote, T. R. Harrison and C. O. Fairchild. (Presented by Mr. Foote.)

Potentiometers for Thermoelement Work. By W. P. White. (Presented by title.)

Self-checking Galvanometer Pyrometer. By H. F. Porter. (Presented by the author.)

Some Factors Affecting the Usefulness of Base-metal Thermocouples. By O. L. Kowalke. (Presented by the author.)

Tables and Curves for Use in Measuring Temperatures with Thermocouples. By L. H. Adams. (Presented by R. B. Sosman.) Illustrated by lantern slides.

Reference Standard for Base-metal Thermocouples. By N. E. Bonn. (Presented by title.)

Alloys Suitable for Thermocouples and Base-metal Thermoelectric Practice. By J. M Lohr. (Presented by title.)

Recent Improvements in Pyrometry. By R. P. Brown. (Presented by title.)

Automatic Compensation for Cold-Junction Temperature of Thermocouple Pyrometers. By F. Wunsch. (Presented by title.)

Hot Wire Anemometer with Thermocouple. By T. S. Taylor. Illustrated by lantern slides. (Presented by the author.)

Porcelain for Pyrometric Purposes. By F. H. Riddle. (Presented by the author.)

Pyrometer Porcelains and Refractories. By R. W. Newcomb. (Presented by title.)

Pyrometer Protection Tubes. By Otis Hutchins. (Presented by title.)

Porcelain Pyrometer Protecting Tubes. By F. A. Harvey. (Presented by the author.) Illustrated by lantern slides.

Protecting Tubes for Thermocouples. By R. B. Lincoln. (Presented by title.)

Melting Point of Refractory Materials. By Leo I. Dana. (Presented by title.)

High Temperature Scale and its Application in the Measurement of True, Brightness and Color Temperature. By Edward P. Hyde. (Presented by the author.)

Theory and Accuracy in Optical Pyrometry with Particular Reference to the Disappearing-filament Type. By W. E. Forsythe. (Presented by the author.) Illustrated by lantern slides.

Optical and Radiation Pyrometry. By Paul D. Foote and C. O. Fairchild. (Presented by Mr. Foote.)

Industrial Application of the Disappearing-filament Type of Optical Pyrometers. By F. E. Bash. (Presented by the author.)

Use of the Optical Pyrometer for Control of Optical-glass Furnaces. By C. N. Fenner. (Presented by title.)

Emissive Powers and Temperatures of Non-black Bodies. By A. G. Worthing. (Presented by the author.) Illustrated by lantern slides.

Recording Thermocouple Pyrometers. By Leo Behr. (Presented by title.)

Recording Pyrometry. By C. O. Fairchild and Paul D. Foote. (Presented by Mr. Fairchild.)

High-temperature Control. By C. O. Fairchild and Paul D. Foote. (Presented by the authors.)

Resistance Thermometry. By F. W. Robinson. (Presented by E. W. Fisher.)

Tin; an Ideal Pyrometric Material. By E. F. Northrup. (Presented by the author.)

Resistance Thermometry for Industrial Use. By Charles P. Frey. (Presented by title.)

Thermocouple Installation in Annealing Kilns for Optical Glass. By E. D. Williamson and H. S. Roberts. (Presented by R. B. Sosman.)

Annealing of Glass. By A. Q. Tool and J. Valasek. (Presented by title.)

Pyrometry Applied to Bottle-glass Manufacture. By R. L. Frink. (Presented by title.)

Pyrometry in the Manufacture of Optical Glass. By Albert J. Walcott. (Presented by the author.)

Pyrometry as Applied to the Manufacture of Optical Glass. By Carl W. Keuffel. (Presented by title.)

Pyrometer Shortcomings in Glass-house Practice. By W. M. Clark and Charles D. Spencer. (Presented by Mr. Spencer.)

Pyrometry in Manufacture of Clay Wares. By F. K. Pence. (Presented by title.)

Application of Pyrometry to the Manufacture of Gas-mask Carbon. By K. Marsh. (Presented by the author.) Illustrated by lantern slides.

Pyrometry in the Ceramic Industries. By C. B. Thwing. (Presented by title.)

Pyrometry in Rotary Portland Cement Kilns. By Leo I. Dana and C. O. Fairchild. (Presented by Mr. Fairchild.)

Pyrometry in the Ceramic Industry. By John P. Goheen. (Presented by Mr. Kellar.)

Temperatures of Incandescent Lamp Filaments. By Benjamin E. Shackelford. (Presented by the author.) Illustrated by lantern slides.

Temperature Measurements of Incandescent Gas Mantles. By H. E. Ives. (Presented by title.)

Applications of Pyrometry to Problems of Lamp Design and Performance. By I. H. Van Horn. (Presented by the author.) Illustrated by lantern slides.

Use of Modified Rosenhain Furnace for Thermal Analysis. By H. Scott and J. R. Freeman, Jr. (Presented by Mr. Scott.)

High Temperature Thermometers. By R. M. Wilhelm. (Presented by title.)

Temperature of a Burning Cigar. By T. S. Sligh and H. R. Kraybill. (Presented by P. D. Foote.)

Teaching Pyrometry in our Technical Schools. By George V. Wendell. (Presented by title.)

Teaching Pyrometry in Technical Schools. By C. E. Mendenhall. (Presented by title.)

Teaching Pyrometry. By O. L. Kowalke. (Presented by the author.)

Present Status of Radiation Constants. By W. W. Coblentz. (Presented by Mr. Wood.)

PAPERS

Blast-furnace Refractories*

BY RAYMOND M. HOWE,† PITTSBURGH, PA.

(Chicago Meeting, September, 1919)

SOME time ago, a prominent engineer asked a representative of the firebrick industry to prepare a comprehensive paper on blast-furnace refractories. It was to have been the purpose of this paper to gather practical experiences from widely different sources, in the hope of determining what kind or kinds of firebrick were best adapted for certain uses. This engineer sent questionnaires to the large consumers of blast-furnace linings and stovebrick and received very detailed replies as to their experiences. These replies are conflicting, but, having been turned over to the writer, form the basis of this attempt to meet the request. Because of the impossibility of making final recommendations at this time, a few of the more interesting and commonly known methods of studying refractories are also discussed, as well as the more recent developments in this field.

The first reply to the questionnaire stated that good service was being secured from the linings. This satisfactory condition was attributed to the use of a very good distributor. The writer believed that manufacturers try to and do produce a good product and that the proper handling of equipment is of the utmost importance, if good results are to be secured. The type of bricks used was not mentioned. The second reply stated that good life was secured from the furnace linings, but that the writer had had soft-fired stovebrick crumble and, for that reason, preferred a hard-fired product for use in this position. The type of bricks used was not mentioned. The third reply stated that firebrick had crumbled from the mantle up; the cause was not known. The fourth reply stated that the stove linings were satisfactory. Hand-made furnace linings failed after 4 years' service so a steam-pressed lining was installed which was still good after 4 years of service. Consequently, a second steam-pressed lining was installed but this disintegrated at the top after the furnace had been in blast 7 months. The firebrick manufacturer attributed this failure to the expansion of iron blocks that had been installed at the stock line to prevent erosion.

The fifth reply stated that the present "hard-burned" hand-made furnace lining had worn away very uniformly to the depth of only 5 in.

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* Appendix will be found on page 761.

† Senior Fellow on Industrial Fellowship of Refractories Manufacturers' Assn., Mellon Institute of Industrial Research.

after 1,042,000 tons of pig had been produced. It strongly advocated the use of only the best clay in laying up linings. No stove trouble had been encountered. The sixth reply stated that the writer had secured 16 and 17 years of service from two sets of stoves. At the end of 17 years, one was torn down because of the poor condition of the shell but the checker work was still good. He was forced to reline the combustion chamber from time to time, but hoped to avoid this by washing the gas. He preferred steam-pressed linings to the hand-made, "as the joints are smaller and scaffolds do not get a hold as easily as on the more open joints."

The seventh reply was more comprehensive and instructive. This writer stated that a number of operating and construction factors beside the physical and chemical nature of the linings themselves affect the life and performance of furnace linings. The most important of these are: Furnace lines, volume of air blown, diameter and lengths of tuyeres, stock line protection, arrangement of gas outlets at furnace top, stock distribution, chemical and physical nature of raw materials, continuity of operation, judgment, knowledge and care by furnace operators. The following data were also given:

Furnace	Blast No.	Days in Blast	Days Operated	Tonnage	Kind of Brick in Lining
2	1	790	744	282,594	Hand-made
2	2	870	870	336,641	Hand-made
2	3	972	972	330,900	Hand-made
2	4	1,607	1,577	628,979	Hand-made
3	1	779	707	252,107	Hand-made
3	2	950	950	365,626	Hand-made
3	3	978	978	351,085	Hand-made
3	4	1,251	1,235	529,316	Steam pressed
4	1	918	904	401,918	Hand-made
4	2	1,664	1,657	823,238	Hand-made

Furnace	Blast No.	Hearth		Bosh		Bosh Angle		Stock Line		Bell		Number of Tuyeres	Kind of top
		Ft.	In.	Ft.	In.	Deg.	Min.	Ft.	In.	Ft.	In.		
2	1	12	6	19	6	72	0	13	6	9	0	12	Handfill
2	2	12	6	20	0	71	20	13	6	9	0	12	Handfill
2	3	12	6	19	6	76	0	13	6	9	0	12	Handfill
2	4	13	6	20	0	74	51	13	6	10	0	12	Handfill
3	1	13	3	20	6	73	18	14	3	10	0	12	Brown
3	2	13	3	20	6	73	18	14	3	10	0	12	Brown
3	3	13	3	20	6	75	0	14	3	10	0	12	Brown
3	4	14	6	21	0	75	25	15	0	10	0	12	Brown
4	1	15	0	22	0	75	0	16	0	11	0	12	Neeland
4	2	16	0	22	0	76	30	16	0	11	9	12	Neeland

He said, "The Brown top on No. 3 furnace during the first three blasts had the original Brown elliptic distributing spout, which had the effect of constantly pitching the stock, so that it was crowded toward the side of the furnace opposite the skip. At the beginning of the fourth blast, the spout was changed to one of V shape, which prevented pitching and effected a wonderful improvement in distribution." He never had brick made to specification but desired a well-burned brick, as pure as possible, made from the proper grinds and mixes to give the necessary physical strength. He prefers a hand-made stove brick to the steam-pressed product, although the latter gives up the heat more rapidly. His preference is due to the fact that his one trial of steam-pressed bricks involved a brick that was too dense, so that spalling occurred.

The eighth reply stated that the user had used the hand-made and the steam-pressed products of one company, but favored the hand-made, although he had never had any blast-furnace lining failures other than the usual erosion. The stove linings had given satisfaction. Tests that had been made on firebrick were also submitted; these are given in Table 1. The ninth reply said the distribution of the stock is the factor that controls the life of a furnace lining, stating "had the same success been achieved toward better blast-furnace tops and proper lines, as has been reached by the refractory brick manufacturers, there would be less said about poor brick." He stated that his company had increased the tonnage secured from three furnaces by changing lines, but was still having trouble with the linings of the fourth, due to no fault of the refractories.

These nine replies, which represented the opinions of those who used enormous quantities of blast-furnace linings and stove brick, were very encouraging to the manufacturers of refractory materials, who had felt at times that perhaps that industry was not meeting the ever-increasing demands being made upon it as higher temperatures are employed. It is interesting to note that four companies secured good service from their furnace linings but gave no reason; one company attributed its success to the distributor being used; another had figures to show how the tonnage secured from each of three furnaces had been increased over 100 per cent. by bettering furnace lines and control. One reply was less complete but was of the same context. The failures are difficult to explain, for the crumbling of a top could have been due to one of three causes, the last of which is improbable because of the conditions under which the furnace operates: The action of furnace gases or carbon;[1] the crushing of the bricks; or the loss of bond after repeated heating and cooling. The influence of furnace gases was recently shown very forcibly when iron was removed from optical-glass pots by chlorine or certain of its compounds. These pots lost all of their bond and fell into pieces.

[1] A review which covers top failures is now being made. This review, together with experimental data, will be presented in the near future.

TABLE 1.—*Properties of Firebrick*

	Chemical Analysis								Physical Tests					
	Kind of Brick	Loss on Ignition, Per Cent.	SiO_2, Per Cent.	Fe_2O_3, Per Cent.	Al_2O_3, Per Cent.	CaO, Per Cent.	MgO, Per Cent.	K_2O, Per Cent.	Crushing Strength, Lb. per Sq. In. Side	Specific Gravity		Porosity, Per Cent. by Volume	Abrasion 10 Lb. for 10 Min. Applied to End	
										True	App.		Wt., Gr.	Cu. In.
1	Top	0.00	52.84	3.95	41.34	0.12	0.21	1.54	....	2.40	2.06	14.08	31	0.060
2	Top	0.07	54.44	2.53	40.01	0.18	0.53	2.24	5098	2.34	2.03	12.93	62	0.120
3	Inwall	0.08	50.20	3.58	43.14	0.25	0.38	2.37	3910	2.28	2.00	12.29	38	0.075
4	H. & B.	0.07	50.57	3.87	42.73	0.25	0.34	2.17	5153	2.33	1.98	14.92	90	0.180
5	Inwall	0.00	52.36	3.25	42.65	0.09	0.19	1.46	3500	2.33	2.00	14.09	18	0.036
6	H. & B.	0.02	59.83	3.34	34.82	0.15	0.34	1.50	3468	2.38	2.07	12.92	252	0.480
7	Top	0.02	64.01	3.08	30.61	0.12	0.36	1.80	3906	2.32	2.01	13.32	25	0.049
8	Inwall	0.04	53.28	3.51	41.00	0.12	0.28	1.77	3750	2.33	2.04	12.71	185	0.360
9	H. & B.	0.02	53.34	3.51	41.18	0.12	0.26	1.57	2711	2.39	1.97	17.54	74	0.150
10	Top	0.00	57.62	3.69	37.62	0.15	0.16	0.76	5360	2.48	2.12	14.57	19	0.035
11	Inwall	0.01	56.07	3.32	39.00	0.12	0.18	1.30	5248	2.43	2.10	13.30	38	0.072
12	H. & B.	0.15	53.19	3.34	40.91	0.56	0.26	1.59	3923	2.52	2.01	19.97	73	0.140
13	H. & B.	0.14	49.66	2.66	45.66	0.31	0.47	1.10	5110	2.46	2.11	13.91	518	0.97
14	H. & B.	0.21	56.30	3.75	37.21	0.37	0.35	1.81	4230	2.48	2.07	16.42	265	0.57

	Furnace Tests					Remarks		
	Average Temp.		Maximum Temp., Deg. F.	Expansion In. per Linear Feet	Average Permanent Elongation per 9 In.	Surface Fusion	Cracking	Fitting due to Iron Oxide
	Deg. F.	Time, Hr.						
1	2960	4	3011	0.052	0.007	Medium	Very slight	Slight
2	3016	2½	3157	0.042	0.021	Slight	Very slight	
3	2960	4	3011	0.042	0.000	Slight	None	
4	2925	3	2983	0.042	0.042	None	None	Very slight
5	....	...	3000	0.075	.....	Very slight	None	Slight
6	....	...	3000	0.064	.....		None	
7	....	...	3000	0.084	.....	Medium	None	Slight
8	....	...	3000	0.073	.....	Medium	Slight	Slight
9	....	...	3000	0.064	.....	Slight	Few and small	None
10	....	...	2905	0.096	.....	Slight	Many small cracks	Badly pitted
11	....	...	3000	0.064	.....	Very slight	None	Very slight
12	2925	3	2983	0.083	0.017	None	None	Slight
13	....	...	2905	0.136	.....	Very slight	None	Very little
14	2983	3	3045	0.034	0.070	Slight	Very little	None

The ninth company secured such varying results from different bricks that it would be useless to attempt any explanation.

The service secured from stove brick appeared to be satisfactory in but two cases: one where the hand-made brick were supposedly underfired and the other where the steam-pressed brick were too dense. For the time being these explanations will be accepted.

The writer then visited different plants and learned other facts concerning failures and successes, all of which tend to show that the opinions of operators as well as manufacturers vary. In one case, a record furnace

lining had been installed with no little concern, because it was so exceptionally hard-burned. As the lining was needlessly repaired, for the first time, after 1,042,000 tons of pig-iron had been produced, this leads one to believe that manufacturers should have no future trouble in disposing of extremely well-burned linings at that plant.

The stove service at the same plant was extremely good, and was attributed to the method of burning the gas. Sufficient air was introduced to burn all the gas in the lower combustion chamber, while other operators allow the gas to burn in the upper combustion chamber and checkers. By following the latter practice, the dust separated out in the relatively cool zone at the bottom of the chamber and, not being fused, was easily removed. Here are two distinct practices. In their adoption it is necessary to consider whether the removal of the dust as dust, or longer service in the top courses of the stoves, is more desirable, for both conditions can hardly exist. At another plant, spalling constantly occurred.[2] The brick proved to be a high-grade soft-clay product. The writer believes that the introduction of grog or flint clay into such bricks, when used in the combustion chamber and the top courses of checkers, would decrease the spalling. The following figures show the effect of the addition of small amounts of flint clay to plastic, the same kind of plastic clay and flint clay being used in each brick.

Experimental Data showing Relative Resistance to Spalling

100 Per Cent. Plastic Clay	90 Per Cent. Plastic Clay 10 Per Cent. Flint	80 Per Cent. Plastic Clay 20 Per Cent. Flint	70 Per Cent. Plastic Clay 30 Per Cent. Flint	60 Per Cent. Plastic Clay 40 Per Cent. Flint
3	4	3[a]	6	22

[a] Erratic result.

At another plant furnaces were operating which had given the following tonnages: 727,330; 467,385; 714,114; 937,359; 391,660; 506,403; 669,740; 609,300. The superintendent attributed his success to the fact that he used cooling plates so freely that he was often criticized. By so doing he secured uniform wear at the bosh rather than the uneven wear resulting from alternate hot and cold spots.

Four cases of brick disintegration were mentioned at different plants. In one case the failure occurred about 30 years ago in one of the old-type furnaces. In the second case, trouble began to occur in a very short time with the furnace using a high-barium high-sulfur ore, and complete failure resulted in a month. In a third case, a furnace was blown out and appeared to be in good condition; it developed, however, upon following a crack, that the inner 5 in. (12.7 cm.) of lining were good, but that the back portions could be crumbled in the hand. In the fourth case the

[2] This question is being further studied.

13½ in. (34 cm.) brick, upon removal, were found to be split into from four to eight sections. This case is entirely different from the others, for those bricks crumbled into a thousand pieces, while these cracked into several pieces, each of which was strong. At another plant, extreme variations were noticed in a lining that was being installed, for the large blocks varied as much as 1 in. (2.5 cm.) in length and were warped in many cases. Two operators who had experience with such blocks thought that unless the blocks could be made with more uniformity they would revert to the smaller brick for furnace bottoms, although theoretically the larger blocks—because of the fewer joints—are better.

Judging by the success of one plant, which more than doubled its tonnage from each lining, requiring less than twice the time to secure such tonnage, it appears as if there is a reason, other than the linings, why erratic results are secured in other places. This increased production was also one of gradual development, which excludes the element of chance. However, in order to secure more uniform results, it is essential that the furnace operator have uniform materials with which to begin. Having these, it is more possible to begin real constructive work. The purpose of this paper is not to tell the steel manufacturers their mistakes or how to control their furnace practice, it will be restricted to showing what the refractories manufacturers are doing to improve their products and make them as uniform as possible. By following such methods, on both sides, it is quite possible that answers from a questionnaire circulated 10 years from now would be free from contradictory statements and might give sufficient information to explain successes and failures in this field.

The manufacturers of firebrick have shown their desire to produce better products from the available raw materials through the action of the Refractories Manufacturers Assn. This association is spending annually a considerable sum of money at the Mellon Institute of Industrial Research in supporting a staff of ceramic engineers and chemists, which has increased in number from one to five in less than 2 years. All of these men devote their full time to the study and improvement of refractory products. In addition, this association has coöperated with the Gas Institute for several years in trying to meet the demands of its members. It has also collaborated with the Bureau of Standards, the Bureau of Mines, and the American Society for Testing Materials in the development of such tests as will determine the most desirable products for their different uses.

Factors Controlling the Quality of Firebrick

The more important factors that influence the quality of a firebrick may be conveniently divided into seven groups. The clay is mined by underground or surface methods and is given a preliminary sorting at the mine

or quarry. It is then transported to the plant, where it is again sorted as the larger pieces are crushed. After crushing to such size as may be handled by the grinding machinery, the clays are either ground separately and then mixed, or they are mixed and then ground. Both the dry and the wet grinding practices are followed. Water is added during the grinding and mixing, the amount depending on the process of manufacture. Then the shaping process follows in which the mix is worked by hand, in a dry press, through a die by different sources of pressure, or in a soft mud machine. After being shaped, the "green" ware is dried in tunnel driers or on hot floors until it is ready to be set in the kiln, burned, sorted, and shipped. The seven steps in the manufacture of firebrick that may affect its quality or appearance, and which are being studied by different firms, are as follows: Mining and sorting of clay, grinding, proportioning, tempering, shaping, drying, burning.

Mining and Sorting of Clay

In order to mine and sort clays on a scientific basis, it is necessary to study each vein from the top to the bottom. This is particularly true of plastic clay, which sometimes shows variations in its different layers; yet when these are known they may be located by their color and structure. Accordingly, the practice of sampling clay is gradually increasing, samples being taken from the operation from day to day or from drill cores obtained years before the clay is used. As a result of such studies, the refractoriness of some brands has been increased by 100° F., while in other cases clay that was not being mined, because of a peculiar color or structure, has proved to be superior to the run-of-mine product. These statements are not meant to imply that all plastic clays vary or that all plastic clays are carefully picked; some plastic clays are remarkably uniform and hence offer no problem, in others, a poor spot may occur in the center of the vein and cannot be entirely removed. Under such conditions the clay can only be carefully mixed into a uniform blend of slightly lower refractoriness, thus avoiding a particularly bad lot. Fusion tests generally give sufficient information as to the uniformity of a deposit. There are times, however, when a particularly siliceous or aluminous brick is desired and this necessitates chemical analyses.

Grinding

The influence of grinding on the product is rather well known; in general, the finer grinds result in the production of a dense firebrick while the coarser grinds produce one that is more open and porous. However, as in the case of concrete, the strongest block is not secured from the sand nor from the gravel, but by having the proper proportions of the different sized material. Fortunately, flint clay or grog assumes different

sizes in grinding and a suitable blend is secured without great difficulty. As a precaution, however, some plants screen their grog to different sizes before incorporating it into the mix. The effect of grinding is also modified in the process of manufacture.

Proportioning

Proportioning is a problem that each plant must work out for itself as each plastic clay varies, in that some will bind considerable grog or flint clay while others will not. Some plastic clays are very nearly as refractory as the materials they bind, while others are of lower refractoriness. Obviously, the permissible amounts of two such clays would not be the same. Another problem that confronts only the producer lies in the relative amounts of plastic flint clay and grog that are available, for he must proportion them in such a way that the maximum return is secured, unless he supplies some one particular trade for which a certain definite type of firebrick is required.

Notwithstanding the fact that 100 per cent. flint-clay bricks are sometimes made, the strength of a brick varies directly with the bonding clay, for the strength of all-flint bricks may be increased by the addition of plastic clay. A direct application of this principle lies in the manufacture of blast-furnace bricks, where the refractory requirements diminish and the strength requirements increase from the hearth to the top.

Tempering

The characteristics of a firebrick depend very largely on the amount of water used in its manufacture. The full plasticity of a clay is not developed unless sufficient water is added; and, conversely, the plasticity may be decreased by diminishing the amount of water used in its tempering. The proper amount of water that should be used in making the densest or most porous brick varies not only with each individual mix but with each process of manufacture. The following figures, selected from several sets of data, illustrate clearly how, in general, the strength of each mix increases with the amount of water until a limit is reached, after which the further addition of water results in a slight falling off in mechanical strength.

Per Cent. Water by Weight	Modulus of Rupture of Burned Piece	Porosity of Burned Piece
7.5	2491	24.0
9.0	6019	20.3
10.0	4020	20.6
	Another mixture of the same clays	
7.5	1891	23.35
9.0	5675	21.45
10.0	4670	21.50

Shaping

Regarding this step, from the standpoint of firebrick structure only, the most dense product should result from clays worked at stiff mud consistency, providing the necessary pressure is applied. Furthermore, a dry-pressed or hand-made brick must be fired to a higher temperature in order to secure the same effect. By the same reasoning, one would hardly choose the stiff-mud process for making an open firebrick.

Drying

Although this part of firebrick manufacture is a study in itself, it is of no particular interest to the consumer. It is true that cracks sometimes develop during the drying of large shapes, and these are often objectionable. It is also true that the color of firebrick is influenced by the water that remains in the brick as they are placed in the kiln, for the combination of sulfur from the coal and steam tend to bring the iron to the surface, giving rise to a pink discoloration. Advantage is taken of this fact in some localities, where it is necessary to set facebrick when they are quite "green," to obtain a sufficiently deep color. Tests have shown, however, that the pink-colored firebrick are often of the same refractoriness as the lighter products made from the same clays, although this color may be an indication of excessive impurities.

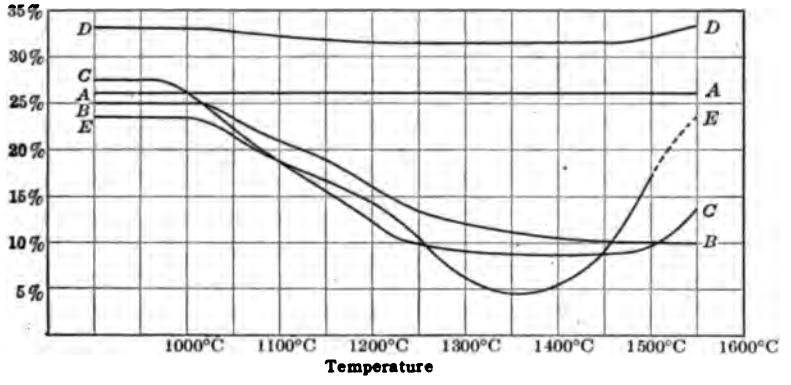

FIG. 1.—POROSITY OF GROUND FIRECLAY PIECES WHEN BURNED AT DIFFERENT TEMPERATURES. *A*, OPEN-BURNING FLINT CLAY; *B*, DENSE-BURNING FLINT CLAY; *C*, HIGH-GRADE DENSE-BURNING PLASTIC CLAY; *D*, HIGH-GRADE OPEN-BURNING PLASTIC CLAY; *E*, LOW-GRADE DENSE-BURNING PLASTIC CLAY.

Burning

Burning is of prime importance to the consumer, who sometimes prefers a light-burned product, while in other instances he wishes one of a very hard burn. The writer does not feel that rule-of-thumb reasoning should be applied to this requirement and submits the curves given in

Fig. 1 to show the effect of heat on different flint and plastic clays. To the manufacturer such curves are important in that they show which clays can be most successfully used in making dense or open firebrick for different purposes; and, while they cannot be expected to be made available to the different consumers, they do indicate what can be expected of the firebricks made from such clays when placed in service. They also act as a guide to the manufacturer in establishing his burning practice, for hardly any two clays require the same treatment in this respect.

When it is borne in mind that clays contract as their porosity decreases, during burning, the following deductions can be quickly and accurately drawn. A flint clay that has a burning behavior similar to *A* need not be burned to a very high temperature in order to make a high-grade product. Its volume remains practically the same and when the bonding clay is mature few, if any, benefits are gained by harder burning. Such a clay is undoubtedly responsible for the statement that certain bricks expand in service and hence give very good service in roofs. As a matter of fact, they can hardly expand more than usual but they do not contract.

A flint clay of type *B*, on the other hand, must be burned in the kiln to a high temperature if shrinkage during service is undesirable. The same statements also hold true in a slightly different way for plastic clays. For example, a plastic clay of type *C* becomes dense at a low temperature and then remains very much the same over a wide range of temperature. Its full strength can be gained and its contraction removed with no difficulty.

On the other hand, a plastic clay of type *E* must be burned to a higher temperature in order to remove its shrinkage, for such a clay continues to contract with increasing temperatures until the period of expansion resulting from overburning begins. A clay of type *D* is not at all uncommon. These plastic clays always retain their open structure and are not adapted to the manufacture of dense, abrasion-resisting bricks. Additional curves could be shown, but these are sufficient to show that all firebricks cannot, or should not, be burned in the same way. They also show how hopeless a task it is to apply any one set of clays to all kinds of service.

INSTALLATION

Were it not for the interest that the manufacturer of refractories must keep in his product until it is torn out, as a success or failure, his responsibility would cease at the plant. It happens, however, that furnace practice and conditions vary so tremendously that repeated failures force him to "see" his product through to the end. A good many failures are simply accepted as such and are not investigated at all, others only superficially, and others are too complicated for solution.

A few failures that have come to the writer's attention, and are not the manufacturer's fault, are briefly described here.

1. A high-grade firebrick fusing at a temperature exceeding 3100° F. (1700° C.) was laid up in a local loam that fused in the vicinity of 2100° F. (1148° C.). Parallel cases concern very cheap unsorted fireclay.

2. A similar firebrick was laid up with thick joints of a patent cement intended for another purpose but, due to the salesman's ignorance or overconfidence, was tried although it was not sufficiently refractory (fusion point 2300° F., 1260° C.) for that purpose.

3. A firebrick used in a furnace roof gave 2 weeks' service where 6 weeks' service was generally secured. Investigation of plant record showed that the coal at that time was extremely high in sulfur and afforded a very corrosive ash, which, when deposited on the roof, formed a slag that ate the brick. Tests made on the good ends of the failure proved the firebrick to be of the usual standard of refractoriness.

4. Lime came into contact with a fireclay brick, several of which in the hot zone reacted with the lime to form a glass. Investigation of the ends proved the firebrick to be of the same quality of refractoriness that had previously given satisfaction. As there was no method for measuring temperature at this plant, the natural assumption was that the heat had become excessive and caused the usual reaction between fireclay and lime or magnesite. Lime is sometimes added to fireclay in order to secure a firm joint; it is difficult to understand why this is done, when manufacturers spend thousands of dollars each year to eliminate high-lime clays and secure those that are low in lime and other similar fluxes.

5. Stove bricks failed, due to a crumbling of the upper courses. The original bricks were of high refractoriness, but tests showed that a very corrosive dust had been deposited which lowered their refractoriness from about 3150° F. to 2400° F. Undoubtedly this dust, as it became heated and cooled or liquid and solid, exerted considerable disintegrating force. It is quite possible that one of the failures mentioned in the questionnaire replies was due to similar conditions.

6. Checker brick were reported as failures. Investigation showed that the bricklayer did not like their color, though fusion tests showed that they were of exactly the same refractoriness as the ones considered satisfactory. The tests also showed that a slag was deposited upon the surface of these bricks, which lowered their softening point of 3175° F. (1747° C.) on the inside of the brick to 2540° F. (1394° C.) on the exposed surface; also that the brick that had been condemned was more dense and, hence, more resistant to slag penetration.

At numerous plants, failures have occurred when sand, ashes, and other forms of insulation have been used. It must be remembered that radiation is often the one factor that saves a firebrick from quick failure. In removing this protection one should proceed with care, the factor

of safety is being decreased and the firebrick is being subjected to more severe conditions, which may result in failure.

On the other hand, roofs sometimes "creep" or sag when made of inferior material that has been too lightly burned. Excessive erosion occurs when a firebrick is not sufficiently strong mechanically to resist this influence. Poorly molded, irregular blocks are also encountered.

It is very rare that a brick melts out unless it is contaminated with furnace dust or slag. Many experiences like the preceding could be mentioned were it necessary, but the one important lesson they teach is that they are costly and do not always give definite information. In order to avoid needless repetition, the manufacturer must know what kind of a product he is producing and the consumer must be thoroughly familiar with his working conditions. The manufacturer is trying to do his part and the consumer must do the same, for he will benefit the most. He must insist on the construction of furnaces according to blueprints and on the careful selection of mortar. He should also know what kind of brick are being installed, not only as to brand but as to burn, process of manufacture, and general characteristics. He should also be able to show temperature and tonnage records for different furnaces and the nature of the fuel or slag, or both, that will be in contact with the firebrick.

These statements do not apply only to blast-furnace linings and stoves, for being major installations they are watched more carefully. The following figures, given before the Federal Trade Commission, show that but a small percentage of the production of firebrick is used in such installations and, hence, if complete records are to be secured they should apply to the other positions. If better records are kept and the causes of failure are noted, there is no reason why many consumers cannot secure the 100 per cent. better service secured at the one plant previously mentioned.

Consumption of Firebrick in Different Industries

	Per Cent.		Per Cent.
Beehives and long coke ovens	0.4	Piping	4.0
Blast furnaces	5.0	Gas plants	4.0
Boilers and stokers	4.0	Glass plants	1.0
Byproduct coke ovens	11.0	Lime plants	2.0
Cement plants	1.0	Malleable furnaces	4.0
Copper, brass, lead, and zinc	2.0	Miscellaneous furnaces	2.0
Crucible furnaces	1.0	Open-hearth plants	41.0
Cupolas and foundries	1.0	Kilns	4.0
Dealers	2.0	Railroad trade	7.0
Rolling-mill furnaces	7.0		

The manufacturer of firebrick, on the other hand, will be better able to make a product that is best suited for each installation or to state that his product does not meet the necessary requirements. If repeated failures occur at some plants, all parties concerned will be in a better

position to know the cause or causes. The Refractories Manufacturers' Assn. has already had some very pleasant experiences, especially in coöperation with the Gas Institute. Should blast-furnace operators feel the need for similar collaboration, this association is not only willing, but is anxious, to join in such a movement.

DISCUSSION

J. S. Unger,* Pittsburgh, Pa. (written discussion†).—I heartily agree with the author's suggestion, near the close of the paper, that greater uniformity in the making and use of brick is desirable. The effect of the manner of laying the brick, the mortar or cement used, the necessity of standardizing the shapes, the elimination of irregular or poorly designed brick, and the necessity for uniform operating conditions were pointed out by me some years ago.

Many factors influence the quality of firebricks. The following figures obtained from bricks during their making or after completion show the need for greater uniformity in their quality. These results were obtained from regular commercial bricks, shipped to the consumer. In many cases the samples were from the same car load.

VARIATIONS IN MOISTURE CONTENT OF PUG AS BRICKS WERE BEING MADE

	Per Cent. Moisture
Column of pug from extrusion machine	9.0
Column of pug from extrusion machine taken 10 min. later	11.0
Moisture in pug from pan having finished grinding pug for blast-furnace bricks	8.7
Same pan, new charge 20 min. later	12.0

FINENESS IN TWO PANS GRINDING SAME MATERIAL FOR SAME GRADE OF BRICK

Pan	Per Cent. on ¼ in. Mesh	Per Cent. on ⅛ in. Mesh	Per Cent. on 20 Mesh	Per Cent. on 40 Mesh	Per Cent. through 40 Mesh
A	1.7	4.2	26.0	20.0	48.1
B	0	1.0	21.0	18.0	60.0

SPALLING LOSS AT 1350°, BRICKS FROM DIFFERENT MAKERS

Kind of Brick	Spalling Loss, Per Cent.
1, Machine made	26.9
2, Machine made	11.7
3, Machine made	9.4
4, Hand made	7.3
5, Hand made	3.4

SPALLING LOSS ON CHECKER BRICKS FROM DIFFERENT MAKERS

	Spalling Loss, Per Cent.
Coarse grind	8.4
Fine grind	49.9

*Central Research Bureau.

†Received Oct. 6, 1919.

CRUSHING STRENGTH OF TEN BRICKS FROM SAME CAR CRUSHED AT 1350° C.

	Pounds Per Square Inch
1	662
2	828
3	1013
4	1112
5	1307
6	1343
7	1581
8	1657
9	1683
10	1708

PENETRATION OF MOLTEN BLAST-FURNACE SLAG AT 1350° C., THREE BRICKS OF EACH BRAND

Brick	Penetration, in Square Inch
A, 1	0.10
A, 2	0.50
A, 3	0.70
B, 1	0.10
B, 2	0.20
B, 3	0.50
C, 1	0.40
C, 2	1.20
C, 3	1.50

An inspection by the eye of two carloads of bricks of the same grade from the same maker showed 15.25 per cent. cracked, warped, badly molded, or otherwise defective bricks in one car and 0.30 per cent. in the other. Many other examples could be given but it is believed the preceding figures are sufficient to show the need for a pronounced improvement in the uniformity of blast-furnace bricks.

THE CHAIRMAN (J. W. RICHARDS, South Bethlehem, Pa.).—It is stated that some disastrous effects resulted because the mortar was not sufficiently refractory. That is not the proper criterion of a mortar for use in laying blast-furnace bricks. The material must not only be of proper refractory power but it must not slag the bricks with which it is in contact. Lime is far more refractory than the bricks, yet it would slag them badly if used between them.

F. H. WILLCOX,* Chicago, Ill.—Blast furnaces nowadays usually have about 8 to 12 ft. (2.4 to 3.6 m.) of iron or steel wearing plate below the bell. I have seen not a few instances in the last months where, immediately under the wearing plate or in the upper inwall, the brick was badly worn back. Immediately below the wearing plates is, perhaps, one of the critical points of failure in blast furnaces. There seems to be very little trouble on the hearth and bosh quality, except that the bottom plugs or blocks are usually irregular.

Other points of failure of brick at furnace plants or the points found difficult of maintenance have been in the stove. These points are in the top 5 to 10 ft. of checkers, in the dome, and also in the combustion chamber itself. In the last few years, most of these stove linings have been made of steam-pressed brick.

In the combustion chamber there have been cases in the last year where the alternating heat has caused a very pronounced spall. In

* Secretary, Freyn, Brassert & Co.

some cases the spalling has progressed through 18 in. (45 cm.) on the side of the combustion chamber, directly into the checkers, so that the checkers fall through into the combustion chamber. The failure is entirely due to the spalling of the brick. On the checkers there is the same spalling tendency, and in addition one sometimes finds a blast-furnace gas that carries a very high concentration of alkali dust. When that is deposited on the top of the stove checkers, in some cases it disintegrates the brick. Any suggestions as to the cure of these points of failure would be welcome. I would like to ask whether it is better to use hand-made or steam-pressed brick of inwall quality for combustion chamber and top of checkers. Basing my opinion on the performance of small checker stoves, my inclination is to keep the steam-pressed brick but to obtain a more refractory, resistant, less dense, and more structurally flexible brick by specifying an increased quantity of flint in stove brick, approximating furnace inwall quality. In regard to wear of top or inwall quality firebrick in the furnace itself immediately beneath the stock-line wearing plates, there would seem to be room for some study as to the cause, either on the assumption of abrasion by stock or disintegration by carbon monoxide or both, and the development of a more resistant quality of refractory. With correct distribution and good furnace lines, cracks in the lining are perhaps the starting point for most of the troubles in the lower inwall.

As to irregularity of size, this is persistent; and while possibly not serious in ring walls of furnace and stoves, it is most serious in furnace bottom blocks and in stove checkers. In the bottom block, irregularities that demand excessive clay to make up prevent a rigid job being put in unless the block is chipped or ground. If the job is not put in rigidly, the blocks float up and one gets a deep bottom in operation, danger of breakout, and a huge salamander.

R. M. Howe.—The addition of flint clay or grog would greatly increase the resistance to spalling. However, it must be borne in mind that, when spalling occurs, the brick is generally heated on one face only, whereas in testing five faces of the brick are suddenly cooled. Hence it is not surprising that large pieces are loosened in such a test and that corresponding variations are noted. In actual service, spalling takes place piece by piece, rather gradually, and as a rule rather regularly. The variation in crushing strength has been realized for a long time but has not been considered serious, for fireclay bricks do not crush; they soften and squeeze out of shape due to the combined action of pressure and temperature.

Production of Ferromanganese in the Blast Furnace*

P. H. ROYSTER,† WASHINGTON, D. C.

(New York Meeting, February, 1919)

ON THE Continent, ferromanganese has been produced in the blast furnace almost continuously since 1876, but little definite information concerning the practice is to be found in technical literature, and almost nothing in the way of a theory of the manganese furnace. In this country, prior to 1904, but one company had attempted its production in a serious way, and none of the operating data or other information from this company has been, or is, available to the public. In the last 5 years, as a result of the restriction of imports, a number of furnaces have been blown in on this alloy, with the result that in 1918 as many as 18 furnaces were in blast. The majority of the men in charge of these furnaces had had little previous experience in such operations; so the Bureau of Mines, in the summer of 1918, undertook the collection of data from ferromanganese furnaces.

Operating data were collected by the Bureau from eleven ferromanganese furnaces, which will be referred to throughout this paper as furnaces A, B, C, F, G, I, J, K, L, M, and P; the dimensions of seven of the furnaces are given in Table 1. The companies from whom this information was obtained, and to whom should go the credit for whatever value this information may possess, are the Bethlehem Steel Co.,

TABLE 1.—*Dimensions of Ferromanganese Furnaces Investigated*

Furnace	Height, Stock Line to Center Line of Tuyeres, Feet	Bosh Diameter, Feet	Hearth Diameter, Feet
A.....	62.1	13.4	8.8
B.....	55.0	16.5	9.5
C.....	66.2	14.8	9.5
F.....	57.8	15.2	10.5
G.....	63.0	18.0	11.5
J.....	67.4	18.3	13.0
K.....	63.0	18.6	13.2

* Published by permission of the Director of the U. S. Bureau of Mines. Abstract of report of research under the joint auspices of the U. S. Bureau of Mines and the National Research Council.

† Assistant Physicist, U. S. Bureau of Mines.

B. & B. Trading Co., Buffalo Union Furnace Co., Donner Steel Co., John B. Guernsey & Co. Inc., E. E. Marshall, Miami Metals Co., Seaboard Steel & Manganese Corpn., Southeastern Iron Corpn., and the Wharton Steel Co.

The essential operating data are given in Table 2, presented in the form of 40 "runs" or experimental periods, each run being the average figures for 10 days continuous operation. In addition to the figures taken from the furnace records, samples of downcomer gas were taken by the Bureau at five furnaces, and analyzed for O_2, CO_2, and CO. At six furnaces, supplementary temperature measurements were taken with Leeds & Northrup pyrometers sighted: (1) down a tuyere, with and without a tuyere glass interposed; (2) on the surface of the slag at flush and at cast; and (3) on the surface of the metal at cast. When the apparent temperatures are corrected, in the case of tuyere temperatures, for the absorption of the glass screen when present and, in the case of the slag and metal, for their respective emissivities,[1] a fair idea of the temperatures existing in the combustion zone, in the slag bath, and in the metal bath of the furnace should be given. Since most of the data in Table 2 concern furnace operation carried out before the visit of the Bureau's field party, it is doubtful whether the gas analyses and temperature measurements taken by the Bureau are strictly applicable to the average practice indicated in Table 2, but no better figures being available, these must serve.

Average Practice for Ferromanganese and Iron Furnaces

Probably the most satisfactory method of discussing the ferromanganese furnace will be by comparing its figures to those of the iron furnace. A comparison of the more important figures for the two practices is given in Table 3. The important differences noticeable are that the ferromanganese furnace received about three times as much coke and over twice as much stone as the iron furnace, lost over five times as much carbon by CO_2 absorption, operated with a hotter top and a colder combustion zone, produced colder metal and slag, and produced gas with a CO/CO_2 ratio two and a half times as great. The points of similarity are that the blast temperatures were roughly the same, and the slag compositions were essentially the same if the dilution of the ferromanganese slag with MnO is ignored. In addition, there are differences due to the nature of the materials used and of the metals to be produced. All of the iron charged that is not blown from the top of the furnace as dust, amounting apparently to 9.6 per cent.,[2] is recovered

[1] The emissivity corrections used are those given by Burgess.

[2] In addition to the Fe in the ore given in Table 3, the iron furnace received an average of 176 lb. (79.8 kg.) of mill scale and 63 lb. (28.5 kg.) of scrap per ton of pig. The iron in the slag is negligible; 5 lb. (2.3 kg.).

TABLE 2.—*Operating Data of Ferromanganese Blast Furnaces*

Experiment	Furnace	Ore, Lb. per Ton	Coke, Lb. per Ton	Coal, Lb. per Ton	Stone, Lb. per Ton	Ore Analysis: Mn, Per Cent.	Fe, Per Cent.	SiO_2, Per Cent.	Tons per Day	Slag, Lb. per Ton	Metal, Per Cent. Mn	Metal, Per Cent. Si	Blast Temp., Deg. F.	Top Temp., Deg. F.	Cu. Ft., Per Min.	Slag Analysis: CaO, Per Cent.	MgO, Per Cent.	Al_2O_3, Per Cent.	SiO_2, Per Cent.	Mn, Per Cent.	Carbon, Lb.	Mn Charged, Lb.	Per Cent. to Metal	Per Cent. to Slag	Per Cent. Lost in Stack
1	A	5981	4658	...	1980	41.18	4.54	7.78	45.7	2671	78.2	0.76	1308	935	12,870	27.2	9.7	13.0	27.8	14.6	3810	2462	71.2	15.8	12.9
2	A	5303	5475	...	1780	39.11	6.65	5.64	37.2	2445	70.6	1.47	1268	1025	12,820	24.1	8.9	16.2	28.2	15.6	4440	2075	76.4	18.4	5.2
3	A	5870	6290	...	1820	39.43	7.42	6.33	29.0	2805	71.3	0.92	1196	1049	.....	23.9	8.6	16.2	26.3	18.9	5360	2316	69.0	22.9	8.1
4	A	6000	7133	...	2320	42.01	3.33	8.45	33.3	3035	80.9	1.24	1232	1054	.....	30.1	10.3	13.2	29.8	11.7	5820	2520	72.0	14.0	14.0
5	A	5090	5016	...	1632	42.25	6.07	6.73	35.7	2248	73.5	1.24	1432	998	12,516	26.5	9.6	13.5	28.7	14.9	4060	2149	76.3	15.5	8.0
6	A	5475	5628	...	1806	39.80	11.28	4.34	40.9	2180	69.4	0.73	1279	1147	.....	30.1	10.9	17.0	27.4	9.4	4560	2180	71.2	9.5	19.3
7	A	5720	5595	...	1860	38.45	14.05	2.58	41.5	2070	68.3	0.85	1303	1103	.....	33.1	11.5	15.4	28.3	9.4	4480	2200	71.0	9.4	19.6
8	A	6166	7352	...	2172	36.66	4.60	6.14	29.5	3335	78.3	0.70	1175	1014	.....	38.2	11.8	14.1	25.5	6.9	5992	2260	77.8	10.1	12.0
9	A	6990	7836	...	2112	37.14	3.69	6.65	31.2	3460	80.35	1.86	1150	979	.....	.....	.....	15.7	26.2	5.3	6380	2600	66.8	7.0	23.7
10	B	7560	8070	...	2700	35.75	6.35	15.06	33.8	4630	72.34	0.62	786	.....	.....	.....	.....	11.1	29.9	13.6	7080	2700	60.0	22.5	17.5
11	B	6212	7776	...	2180	36.70	4.41	17.05	35.0	4110	77.13	0.58	770	.....	.....	.....	.....	12.5	28.7	9.5	6810	2280	75.8	17.1	7.1
12	B	7545	8435	...	2475	36.00	6.14	14.97	33.8	4460	72.88	0.74	886	.....	.....	.....	.....	11.0	30.5	13.3	7400	2715	60.1	21.9	18.0
13	B	6025	7534	...	2265	38.00	3.43	15.28	34.7	3670	77.7	1.43	782	.....	.....	44.3	.....	14.1	29.8	9.1	6600	2292	76.0	14.6	9.4
14	B	8102	9510	...	3320	34.00	7.75	17.84	35.0	5745	74.5	0.58	830	.....	.....	35.5	.....	10.7	31.3	15.6	8340	2760	60.6	32.4	7.0
15	F	7099	7539	...	3168	42.91	4.09	11.30	32.0	4100	76.3	0.37	735	757	14,380	28.4	6.5	11.1	29.1	19.8	6220	3045	56.1	25.6	18.3
16	F	4813	6092	...	2091	43.68	6.71	6.87	44.9	2470	79.1	0.83	993	328	22,500	35.0	9.9	10.3	27.5	13.2	5030	2108	84.4	15.5	0.1
17	F	4831	6431	...	2084	45.76	5.90	8.44	44.7	2610	78.8	1.12	1027	143	22,500	33.3	11.2	11.5	27.8	11.3	5310	2218	80.2	13.3	6.5
18	G	7525	5460	546	2941	32.53	5.60	12.62	56.5	4600	78.2	1.53	1140	730	17,500	25.2	10.7	13.9	29.8	11.3	5140	2450	71.5	24.9	3.6
19	G	6050	4863	459	2675	32.52	5.60	12.62	60.7	3765	67.5	1.61	1183	780	16,900	26.2	12.8	15.2	30.9	11.4	4565	1969	77.0	21.8	1.2
20	I	8150	7282	...	3160	39.26	2.40	15.54	50.0	5040?	71.4	1.92	952	1016	17,100	26.6	14.1	11.3	27.7	11.4	5990	3200	52.7	18.0	29.3
21	I	5850	4652	...	2165	41.64	5.54	7.43	82.7	2555	77.1	0.56	1277	934	16,500	31.4	12.6	11.4	26.5	12.0	3825	2435	71.0	12.6	16.4
22	J	8560	6940	...	2510	34.83	4.00?	14.92	60.8	6230	69.1	2.05	1108	849	20,000	24.2	11.1	13.8	30.4	15.1	5550	2985	51.6	31.2	17.2
23	J	5040	500	432	1278	44.73	4.06	2.60	94.1	2050	79.2	0.75	1156	496	19,500	26.0	8.9	24.2	21.7	8.1	4395	2258	78.7	7.4	13.9
24	J	4900	2900	415	1264	45.20	3.80	2.53	70.0	2020	79.7	0.63	1237	....	20,000	27.3	7.1	23.5	21.3	13.3	4340	2226	80.3	12.1	7.6
25	J	4960	4955	419	1255	43.75	4.18	2.86	97.9	1985	79.3	0.61	1172	562	19,500	27.9	7.1	24.5	20.9	12.1	4434	2168	81.6	11.1	7.3
26	K	4040	6275	...	1985	50.65	7.27	6.52	46.9	2190	75.9	2.43	950?	....	.....	46.4	1.7	14.7	27.3	5.4	5420	2048	83.0	5.8	11.2
27	K	4480	5802	...	2268	51.13	7.10	5.35	85.8	2580	74.8	0.29	970	965	17,500	37.1	1.1	11.4	28.8	14.3	4690	2288	73.1	16.2	10.7
28	K	4650	6368	...	2184	50.11	7.30	5.57	49.5	2315	75.9	1.82	950?	....	.....	46.6	1.5	13.8	27.6	5.7	5450	2330	73.0	5.7	21.3
29	K	5215	6329	...	2386	46.94	10.80	6.24	77.5	3285	62.5	0.26	998	845	19,100	33.3	1.4	11.1	26.5	19.6	5280	2442	57.2	26.3	16.5
30	L	6770	7458	660	2965	39.97	3.51	7.53	48.1	3126	80.7	1.29	1065	1162	17,800	40.3	6.8	11.7	31.1	8.8	6730	2710	66.9	10.3	22.8
31	L	6565	8459	747	3100	39.68	3.69	7.40	43.0	3305	81.1	1.47	1043	1152	17,700	47.7	5.7	12.8	27.4	4.6	7655	2608	69.7	5.9	24.4
32	L	5312	6500	524	2582	35.45	9.54	8.91	64.7	2886	69.8	0.75	1089	1162	18,800	42.0	5.6	11.8	29.9	7.2	5951	1888	82.8	10.1	7.1
33	L	6222	6421	424	2528	35.38	4.50	6.28	50.5	2727	78.4	2.72	1204	1127	17,800	43.3	4.3	14.5	27.3	6.9	5651	2200	79.7	11.1	9.2
34	L	5465	4855	471	2200	32.38	8.02	10.11	77.2	2452	70.4	1.46	1457	1123	18,800	41.9	6.3	10.5	31.6	5.8	4432	1770	89.0	8.1	2.9
35	L	5200	5164	352	2155	33.53	7.81	8.63	77.1	2256	71.8	2.23	1207	1140	18,800	44.3	4.9	13.2	31.0	4.4	4568	1745	92.1	8.7	0.8
36	P	7000	5350	545	3591	38.70	4.25	15.89	61.3	4468	78.1	....	1450?	750?	.....	41.0	....	14.9	35.4	4.5	5060	2718	64.5	6.4	29.1
37	P	4117	4965	506	1680	45.50	5.55	4.85	78.0	1880	80.9	....	1450?	750?	.....	45.6	....	18.8	20.3	9.9	4700	2060	88.0	4.9	7.1
38	C	5245	5704	811	2773	40.70	7.45	7.01	35.0	2700	74.9	2.19	1387	952	.....	29.8	18.2	10.7	32.2	7.3	5280	2138	78.7	8.9	12.3
39	M	6586	7118	998	3042	45.11	3.66	6.59	42.0	3900	75.7	1.20	1242	905	.....	38.8	....	14.4	25.7	15.2	6500	2970	57.1	20.0	22.9
40	M	6598	7100	957	3510	46.36	4.15	4.08	36.8	3480	74.5	1.30	1182	910	.....	43.3	....	11.1	28.7	11.8	6471	3056	54.5	13.4	32.1
Average.		5992	6326	200	2349	40.33	5.93	8.60	51.7	3196	74.9	1.15	1135	690	.....	Bases 41.75		14.0	28.1	10.6	5323	2382	72.0	14.7	12.8

in the metal. Of the manganese charged in the ferromanganese furnace, however, only 72 per cent. finds its way into the metal; about 15 per cent. of it is carried away unreduced in the slag and 13 per cent. is carried out at the top of the furnace with the gas, a loss that will be called, for convenience, the stack loss.

Control of Furnace Losses

There are no silicon specifications to be met in the making of ferromanganese; the sulfur from the coke is taken care of as MnS in the slag, and, due to the large amount of coke in the charge, the stock is "loose," descending easily in the furnace without hanging or slipping. The operating difficulties, therefore, resolve themselves practically into the problem of keeping down the slag and stack losses. According to generally accepted ideas, the slag loss is reduced by operating with a highly basic slag, with a hot blast, and with plenty of coke. According to the

Table 3.—*Comparison of Operating Data for the Ferromanganese Furnace and for the Iron Furnace*

	Ferro-manganese	Pig Iron		Ferro-manganese	Pig Iron
1 Ore per ton metal, lb...	5,992	3,937	22 CO in gas, per cent....	31.00	23.57
2 Coke per ton metal, lb..	6,326	1,969	23 H_2 in gas, per cent....	0.17*	0.17
3 Coal per ton metal, lb..	200	0	24 CH_4 in gas, per cent...	0.21*	0.21
4 Stone per ton metal, lb..	2,349	984	25 N_2 in gas, per cent....	58.36*	56.16
5 SiO_2 in ore, per cent....	8.6	7.55	26 SiO_2 in slag, per cent..	28.1	36.14
6 Al_2O_3 in ore, per cent...	3.0*	1.96	27 Al_2O_3 in slag, per cent.	14.0	13.33
7 Fe in ore, per cent.....	5.9	52.41	28 CaO in slag, per cent..	−41.7	43.59
8 Mn in ore, per cent.....	40.3	0.61	29 MgO in slag, per cent..	−41.7	4.50
9 SiO_2 in coke, per cent...	6.5*	4.98	30 Mn in slag, per cent..	10.6	0.53
10 Al_2O_3 in coke, per cent..	3.0*	3.25	31 S in slag, per cent....	....	1.49
11 F. C. in coke, per cent..	85.5	87.64	32 Slag volume, lb.....	3,196	1,150
12 SiO_2 in stone, per cent..	....	3.33	33 Blast temperature....	1135° F.	1031° F.
13 Al_2O_3 in stone, per cent.	....	1.28	34 Top temperature.....	690° F.	400° F.
14 CaO in stone, per cent..	−52.0	48.20	35 Temperature combustion zone.....	2740° F.	2812° F.
15 MgO in stone, per cent..	−52.0	3.64	36 Temperature of slag...	2515° F.	2740° F.
16 C in metal, per cent....	6.5*	3.80	37 Temperature of metal.	2435° F.	2740° F.*
17 Si in metal, per cent....	1.15	1.33	38 Wind per ton metal, lb.	24,000	8,012
18 Fe in metal, per cent...	17.0*	93.85	39 Gas per ton metal, lb.	31,610	10,634
19 Mn in metal, per cent..	74.9	0.87	40 Metal per day, tons...	51.7	499
20 S in metal, per cent....	0.01*	0.036	41 C burned at tuyeres...	4,165	1,386
21 CO_2 in gas (by weight) per cent.....	10.44	19.86	42 C absorbed by CO_2...	946	169

* Estimated.

generally accepted ideas, the stack loss is supposed to increase whenever the slag loss is reduced, in particular, with the hearth temperature, with the basicity of the slag, and with the temperature of the top gas. The evidence offered by this investigation fails to support any of these theories; fails, in fact, to show any marked relationship between the stack loss and any other probably related quantity.

Slag Loss

The most self-evident statement to be made concerning slag loss, but the one that actually is seldom stated and often is forgotten, is that the amount of manganese carried away by the slag equals the product of the slag volume and the percentage of manganese in the slag. A low percentage of manganese in the slag has grown to be something of a commonly accepted criterion of good practice; where the manganese content of the slag is reduced as a result of high blast temperature or by slow driving, it is a satisfactory guide, but if it is reduced by the addition of coke, especially a high-ash coke, the increase in slag volume may outweigh the advantage of a reduction in the manganese percentage of the slag with the result that the actual weight of manganese in the slag is greater.

The first step in the problem of reducing the slag loss is to find a means of determining in advance the percentage of manganese in the slag, rather than to find by what means it may be made a minimum. The results of the 40 runs were plotted, first with pounds of carbon charged as ordinates and blast temperatures as abscissas, and second, with pounds of carbon charged as ordinates and slag volumes as abscissas. Through the points the two best straight lines were drawn, and from the slopes of these lines the variation in carbon fuel with blast temperature and with slag volume was determined. It was found that:

$$\left.\frac{dK}{dV}\right|_{T = \text{const.}} = 0.88$$

and that

$$\left.\frac{dK}{dT}\right|_{V = \text{const.}} = -1.33$$

in which T = blast temperature, in degrees F.;
V = slag volume, in pounds per ton of metal;
K = pounds of carbon in fuel per ton of metal.

Integrating these two formulas and determining the constant of integration

$$K_a = 4170 + 0.88V - 1.33T \qquad (1)$$

in which K_a is carbon required with given blast temperature and given slag to maintain average temperature conditions in hearth. That is, if for every degree increase in blast temperature the carbon is reduced 1.33 lb. (0.6 kg.) and for every pound increase in slag volume the carbon is increased 0.88 lb. (0.4 kg.), the hearth temperature conditions will re-

main constant.[3] It does not mean that the operator must increase the carbon charged 0.88 lb. for every pound increase in slag volume. Since usually the man burdening the furnace does not know what the slag volume is and does not admit that the figure 0.88 is correct, such a contention would be absurd. If he wishes, the operator can reduce the carbon charged until the hearth almost freezes or he can charge so much carbon that he bankrupts the owners, and the furnace will still make ferromanganese, even if it does not make money.

The figures in Table 2 were arranged, first, in order of decreasing "excess carbon" $(K - K_a)$; second, in order of decreasing "speed of slag formation," and third, in order of decreasing slag basicity $(CaO + MgO)/SiO_2$. Each of these series was broken into four groups and the various figures averaged by groups. The results appear in Tables 4, 5, and 6. The significance of the first five columns is self-evident. The manganese in the slag is seen to be a function of three variables. In order to determine the effect of a change in one of these variables while the other two remain constant, it was necessary to apply the method of successive approximation. This somewhat tedious method shows that:

$$\frac{dM}{d(K - K_a)} = 0.0011, \text{ if } B \text{ and } R \text{ are constant}$$

$$\frac{dM}{dR} = 1.34, \text{ if } B \text{ and } (K - K_a) \text{ are constant}$$

$$\frac{dM}{dB} = -9.16, \text{ if } R \text{ and } (K - K_a) \text{ are constant}$$

in which M = percentage of manganese in slag;
B = ratio of $(CaO + MgO)/SiO_2$ in slag;
R = pounds of gross slag made per minute per square foot hearth area:[4]

Integrating the three formulas and determining the constants of integration.

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[3] K_a and K are but two values of the same variable. K_a is not an operating quantity of the furnace, it is the carbon fuel that the furnace would receive if it were given the average fuel relative to its slag volume and blast temperature. If the actual fuel charged is more than K_a, it will cause the hearth to operate at a higher temperature; if it is less than K_a, the hearth will be at a lower temperature than the average. In fact $(K - K_a)$ is an arbitrary temperature scale, although it is not possible to connect its values with the thermometric scale.

[4] Gross slag means the total weight of slag-forming materials, manganese being a slag-forming material. The evidence at hand points strongly to the conclusion that the MnO is reduced in the hearth. Therefore the amount of this gross slag—a basic manganese aluminosilicate—formed per minute per square foot of hearth area is the best measure of the speed of operation. This rate of slag formation is given throughout in the unit of pounds per minute per square foot of hearth area.

$$M = 10.6 - 0.0011\,(K - K_a) + 1.34R - 9.16B \qquad (2)$$

Combining formulas 1 and 2, and multiplying the result by V,

$$\text{Slag loss} = MV = 15.2 + 1.34RV - 0.0011KV + 0.000968V^2 - 0.00146TV - 9.16BV \qquad (3)$$

The meaning of the last column in Tables 4, 5, 6 will now be apparent. In Table 4, the relation between M and $K - K_a$ is to be shown, but the values of B in the third column and R in the fourth column are not constant. The actual change is due largely to the change in $K - K_a$ but

TABLE 4.—*Variation of Per Cent. Mn in Slag with Excess Carbon*

Group of Runs	Excess Carbon, Pounds	Bases / SiO_2 in Slag	Rate of Slag Formation	Mn in Slag, Actual Per Cent.	Mn in Slag, Reduced Per Cent.
1	+754	1.69	1.84	8.2	10.3
2	+354	1.48	1.92	11.0	11.0
3	− 56	1.45	2.18	12.1	11.4
4	−765	1.37	2.36	13.3	11.8

TABLE 5.—*Variation of Per Cent. Mn in Slag with Rate of Slag Formation*

Group of Runs	Rate of Slag Formation	Bases / SiO_2 in Slag	Excess Carbon, Pounds	Mn in Slag, Actual Per Cent.	Mn in Slag, Reduced Per Cent.
5	2.86	1.38	− 60	13.0	13.9
6	2.38	1.70	−312	12.0	12.9
7	2.15	1.51	− 62	12.6	12.5
8	1.60	1.64	+240	10.3	12.2

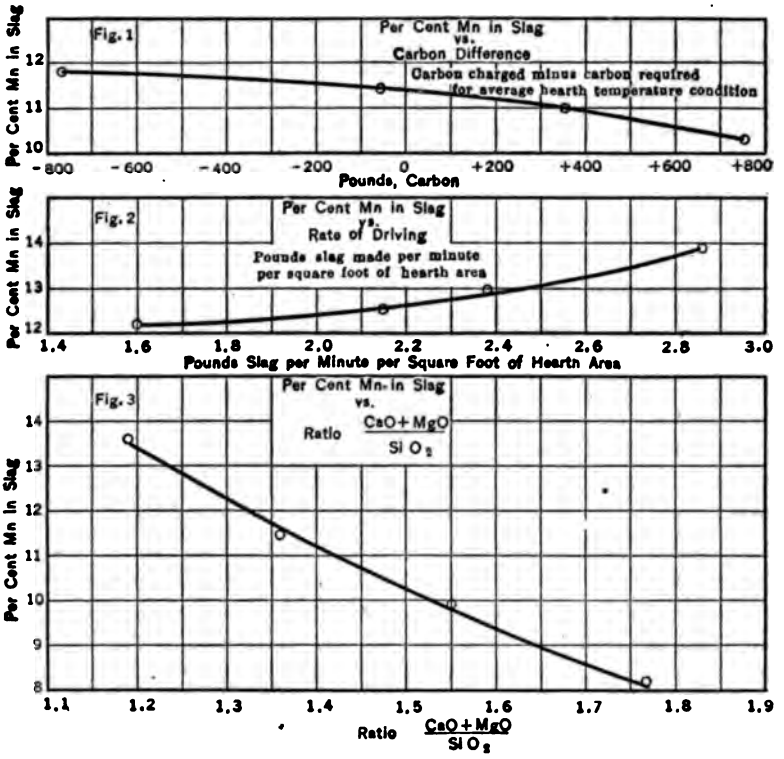
TABLE 6.—*Variation of Per Cent. Mn in Slag with Basicity*

Group of Runs	Bases / SiO_2 in Slag	Excess Carbon, Pounds	Rate of Slag Formation	Mn in Slag, Actual Per Cent.	Mn in Slag, Reduced Per Cent.
9	1.77	+411	1.79	7.4	8.2
10	1.55	+187	2.06	9.2	9.9
11	1.36	− 10	2.32	11.8	11.5
12	1.19	−110	2.42	14.2	13.6

to some extent to changes in B and R. The numbers in the first column of Table 4 are used as abscissas in Fig. 1, and the reduced values of M in the sixth column are used as ordinates. The results from Tables 5 and 6 are reduced in the same manner and plotted in Figs. 2 and 3.

Stack Loss

The 40 runs of Table 2 were divided into two groups according to the values of the stack loss for each run and the averages of these two groups are given in Table 7. Although the difference between the stack losses for the two groups of runs involves 14 per cent. of the total manganese charged, there is no noticeable relationship between it and the other figures given in the table. It should be remembered that the stack loss



Figs. 1, 2, 3.

Table 7.—*Averages of Two Groups Formed from Runs of Table 2*

	Operation with High Stack Loss	Operation with Low Stack Loss		Operation with High Stack Loss	Operation with Low Stack Loss
Stack loss, per cent.....	20.0	6.5	Silicon in metal, per cent	1.19	1.11
Top temperature, degrees			Ratio of bases to silica..	1.54	1.52
F.....	924	930	Excess carbon.....	+83	−20
Mn in slag, per cent.....	11.3	10.9	Rate of slag formation..	2.24	2.09

is the result of a computation involving four weights and ten chemical analyses, and that the nature of the computation is such that any errors in the determination of these fourteen quantities accumulate. In the absence of better information, it must be taken as a working hypothesis that the stack loss is not increased by a hot top, a basic slag, fast driving, or a high coke consumption.

Burdening the Furnace

If this conclusion concerning the stack loss is accepted, formula 3 will be sufficient for determining in advance from the charge sheet the results to be expected from the furnace. Casual examination of Figs. 1, 2, and 3 will make it apparent that the reduction of the percentage of manganese in the slag by using excess carbon, or even by using the average carbon, will hardly pay for itself. It would seem that, in general, more coke has been used than was profitable; also, the rate of driving has been too low, at least from the standpoint of furnace profits. The most effective method of reducing the slag loss is by carrying a highly basic slag. In nearly every case a ratio of bases to silica in the slag as high as 2.0 is practicable. When the alumina content of the ore is high, it is possible to operate with a slag having a basicity of 2.5 or greater; in this case it is possible to bring the percentage of manganese in the slag down to 1.5 or 2.0 per cent.

The saving possible by changes of practice, while appreciable, is very much less than that resulting from the proper choice of materials. It can be shown that, by the use of low-silica coke and stone, the profits of the furnaces could have been doubled, and this with a saving of 9 per cent. of the ore and 38 per cent. of the coke used, and with a 50 per cent. increase in furnace capacity. It has been suggested that great advantages are to be derived from a high blast temperature. According to formula 1, bringing the average temperature up to 1700° F. (927° C.) would lower K_a 750 lb. (340 kg.) of carbon. The saving, however, would undoubtedly be out of proportion to the trouble and expense needed to increase the blast temperature to 1700° F.

Accuracy of Result

The operating data recorded here are, in general, less reliable than would be expected by one acquainted only with iron-furnace practice, for: (1) Manganese ores are extremely variable in character and in chemical composition; (2) most of the manganese-furnace operations included in this investigation were undertaken and carried out at a time when it was impracticable to maintain the clerical force and the chemical laboratory required to keep perfectly satisfactory furnace rec-

ords. It was a time, too, when the coke delivered at the furnace was becoming increasingly variable in composition.

Where relationships have been worked out, however, based on the average figures from a number of different furnaces, there is a tendency for any errors present to neutralize one another. It is reasonable to suppose, therefore, that the results given above are in general correct. That the results, if true, would be worth over $20,000,000 annually to the manganese-furnace industry means only that such an investigation was needed; the fact that they may be untrue means only that a more complete investigation is needed. The writer's purpose in working out these relationships was, in fact, largely to show the nature, extent, and value of the results that can be obtained from an investigation of this kind.

Manufacture of Ferro-alloys in the Electric Furnace

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(Colorado Meeting, September, 1918)

BEFORE the outbreak of the war in 1914, the only electric-furnace smelting plant operating on a commercial basis west of the Mississippi River was an electric pig-iron plant in California; rare metal ores were not smelted, concentrates of molybdenum, tungsten, uranium, and vanadium ores being shipped East for reduction. Practically no chrome ore was being mined, and only comparatively small amounts of manganese ore. Today there are two electric smelting plants in California, one in Washington, one in Iowa, and four in Colorado; two plants are under construction in California, and one in Montana. Ores of manganese, chromium, molybdenum and tungsten are now smelted in the West, and although uranium and vanadium concentrates are still shipped to the East, they will probably be treated in western plants in a short time. The four electric-furnace plants in Colorado are listed in Table 1.

TABLE 1.—*Electric Smelting Plants in Colorado*

COMPANY	LOCATION	CAPACITY, KILOWATTS	PRODUCTS
Tungsten Products Co.....	Boulder	800	Ferrotungsten, ferromolybdenum.
Boulder Tungsten Production Co..	Boulder	400	Ferrotungsten.
Ferro Alloy Co.....	Utah Junction	1200	Ferrochrome, ferrotungsten.
Iron Mountain Alloy Co.....	Utah Junction	3000	Ferromanganese.

Power for all of these plants is supplied by the Colorado Power Co.

The plant of the Ferro Alloy Co., at Utah Junction, contains one 750-kw., three-phase furnace, to which power is supplied by three 250-kw. transformers, connected $\triangle/Y$ to give 129 volts on the furnace cables, the transformer ratio being 13,200 to 75 volts. The furnace operates with an actual voltage of 120 volts, and produces about 3 tons of ferrochrome per 24 hr. from 40-per cent. Cr_2O_3 ore. The furnace consists of a steel shell of circular cross-section, 8 ft. (2.43 m.) in diameter by 7 ft. (2.13 m.) deep, lined with magnesite, and having three vertical carbon electrodes 12 in. (30.48 cm.) in diameter.

A 450-kw. furnace, of the same size as the 750-kw. furnace, is also operated on chrome ore. This furnace is three-phase in appearance, but operates electrically as a single-phase furnace. There are three vertical

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carbon electrodes 8 in. (20.32 cm.) in diameter with a conducting carbon bottom. Power is supplied by three 150-kw., single-phase transformers, two of which have a ratio of 13,200 to 95 volts, and one a ratio of 13,200 to 100 volts. From each transformer, one lead goes to one vertical electrode, and one lead to the carbon bottom, so that the transformers are electrically independent. The furnace voltage varies from 90 to 95.

At intervals, 150-kw., single-phase furnaces are operated on ferrotungsten. These furnaces have one vertical graphite electrode 4 in. (10.16 cm.) in diameter, with a water-cooled steel bottom contact. The furnace shells are 4 ft. (1.22 m.) in diameter, and are mounted on trunnions, the slag being poured by tilting the furnace. All of the electrodes are regulated by hand. The chrome furnaces are operated without roofs and the tungsten furnaces with roofs. The whole plant has a power factor of 90 per cent.

The plant of the Iron Mountain Alloy Co., at Utah Junction, contains one 1200-kw., three-phase furnace, and one 1800-kw., three-phase furnace, giving a total capacity of 12 long tons of ferromanganese per 24 hours.

The 1200-kw. furnace is supplied with power by three 400-kw., single-phase transformers, having a ratio of 13,200 to 75 volts, and connected Δ/Δ. The voltage on the furnace is 72 volts. The 1800-kw. furnace is connected to three 600-kw., single-phase transformers having a ratio of 13,200 to 75 volts, and connected Δ/Δ. The power factor of the plant is 85 per cent.

The furnace shells in both cases are 18 ft. (5.48 m.) long, 8 ft. (2.43 m.) wide, 7 ft. (2.13 m.) deep, and are lined with magnesite. The electrodes on the smaller furnace are of carbon, 17 in. (43.18 cm.) in diameter, and on the large furnace 24 in. (60.96 cm.) in diameter; in both furnaces, they are threaded for continuous feeding. No roofs are used, and regulation of electrodes is by hand.

Ferrochrome

The process of making ferrochrome from domestic chromite ore consists of mixing ore, coal or coke, lime, fluorspar, or silica, in proper proportions, and smelting this mixture in an electric furnace. At 2-hr. intervals, the ferrochrome and slag are tapped into iron pots, the metal settling to the bottom. The pot is dumped when the contents are solidified. Reduction theoretically takes place according to the reaction:

$$FeO.Cr_2O_3 + 4C = Fe.2Cr + 4CO.$$
$$9FeO.9Cr_2O_3 + 50C = 2[(Fe_3C.3Cr_3C_2) + 3Fe] + 36CO.$$

In smelting domestic ore, which has a higher percentage of slag-forming constituents than foreign ore, and averages 40 to 45 per cent. Cr_2O_3, considerable [illegible] loss of chromium occurs and it is more difficult to

control the carbon content of the product. By permitting a higher slag loss, the carbon can be kept below 6 per cent., but varies between 4.5 and 8 per cent. The products are sorted into ferrochrome with 6, 7, and 8 per cent. carbon; the greatest demand is for the 6-per cent. carbon product, but large quantities of the other grades are used by the

TABLE 2.—*California and Oregon Chromite Ores*

	1 Per Cent.	2 Per Cent.	3 Per Cent.	4 Per Cent.	5 Per Cent.	6 Per Cent.	7 Per Cent.	8 Per Cent.	9 Per Cent.	10 Per Cent.	11 Per Cent.	12 Per Cent.
Cr_2O_3	51.40	46.45	42.25	42.50	42.80	50.50	48.10	44.43	41.90	44.08	34.27	41.91
FeO	16.71	13.88	13.37	13.11	14.14	10.54	13.63	15.42	15.42	11.57	11.05	15.42
Al_2O_3	17.26	26.00	13.40	19.66	15.08	19.98	17.54	14.30	16.60	20.80	11.90	16.60
MgO	6.60	7.89	13.02	8.37	8.04	11.00	12.20	13.76	12.82	16.15	21.30	12.82
CaO	Tr.	Tr.	1.00	Tr.	0.80	Tr.	Tr.					
SiO_2	4.26	6.60	8.84	5.48	9.94	4.90	5.30	9.44	8.70	6.90	21.70	8.70
Sulfur	0.093	0.063	0.07	0.44	0.233	0.022	0.037	0.035	0.025	0.030	Tr.	0.025
Phosphorus	0.038	0.02	0.03	0.049	0.065	0.063	0.034	0.043	0.036	0.024	0.035	0.036
Copper	0.06	0.02	0.007	nil	nil	0.09	0.09	0.003	0.004	0.004	0.004	0.004
Arsenic	nil	nil	nil	nil	nil	nil	nil					
Tin	nil	nil	nil	nil	nil	nil	nil					

steel industry. The iron content of domestic ores is higher than that of foreign ores, with the result that the chromium content of the product varies from 60 to 65 per cent. instead of 65 to 70 per cent.

Analyses of typical domestic ores are given in Table 2, the ore all having been produced in California or Oregon. The analyses represent

carload shipments; No. 1 to 7 were produced in 1917, and the rest in 1918; the decreasing grade of the ore is noticeable.

Effect of Grade of Ore, Grade of Product, and Constituents of Charge on Recovery and Carbon Control

In smelting chrome ore to produce a ferrochrome with 8 per cent. carbon, considerably higher recovery can be attained than when making

TABLE 3.—*Production of Ferrochrome with 6 to 8 Per Cent. Carbon*

	1	2	3	4	5	6	7	8	9	10	Average
Charge:											
Chrome ore, lb.	11,890	10,333	11,223	12,211	12,858	13,238	12,533	12,150	12,703	12,355	
Coke, lb.	3,961	3,411	3,643	3,847	4,188	4,071	3,615	3,423	3,063	3,617	
Lime, lb.	132	115	122	105	165	308	394	235	116	465	
Fluorspar, lb.	332	298	300	80	167	192	202	192	151	182	
Slag, 20 per cent. Cr, lb.	2,099	1,821	1,933	2,155	2,268	2,334	3,552	4,088	3,966	4,901	
Ferrochrome:											
Chromium, per cent.	65.20	65.80	65.20	64.50	64.00	63.80	64.10	65.70	64.90	63.2	64.64
Carbon, per cent.	6.75	6.80	6.25	6.47	6.73	6.17	7.67	7.81	7.09	7.77	6.95
Slag:											
Chromium, per cent.	4.20	5.60	3.68	5.75	5.50	4.53	4.30	4.45	9.54	4.53	5.21
Carbon charged per 100 lb. chrome ore, lb.	18.9	18.9	18.4	17.9	18.6	17.6	16.5	16.1	13.8	16.8	17.3
Weight ferrochrome, lb.	5,286	5,139	5,327	6,529	6,909	6,977	6,809	7,039	6,598	7,142	
Weight chromium, lb.	3,430	3,380	3,479	4,210	4,422	4,447	4,378	4,639	4,261	4,508	
Weight slag, lb.	5,261	4,777	4,370	6,038	4,793	6,004	8,157	8,678	8,349	8,497	
Chromium recovered, per cent.	88.5	99.2	95.2	97.7	97.8	95.3	93.4	96.2	85.75	91.7	94.07
Kw.-hr. per lb. ferrochrome	4.36	3.9	3.57	3.68	3.19	3.42	2.50	3.7	3.04	3.65	3.50

Note.—Recovery is based on metal output, and ore and slag charged. The operations were conducted in three 150-kw. furnaces and three 250-kw. furnaces.

a 6-per cent. carbon product, for two reasons: (1) excess of carbon permissible in the charge; (2) when it is not necessary to limit the carbon

to 6 per cent., fluxing materials can be charged, giving a more fluid slag and resulting in better separation of slag from metal. There is no difficulty in obtaining a recovery of 90 to 95 per cent. of the chromium when making 8-per cent. carbon product, while the recovery on the 6-per

TABLE 4.—*Production of Ferrochrome with 4 to 6 Per Cent. Carbon*

	11	12	13	14	15	16	17	18	19	20	Average
Charge:											
Ore, lb.	16,473	14,331	13,226	15,079	15,232	16,572	18,666	20,468	18,768	19,482	
Coke, lb.	4,215	3,661	3,434	3,843	3,941	3,921	3,996	4,214	4,304	4,275	
Lime, lb.	452	390	360	390	424	230	180	29			
Sand, lb.	1,130	975	900	892	1,383	1,834	1,932	2,270	2,000	2,120	
Slag, 16 per cent. Cr., lb.	2,907	2,529	2,334	2,661	2,688	2,924	3,294	3,612	3,282	3,438	
Ferrochrome:											
Chromium, per cent.	64.0	62.8	64.1	62.2	63.2	64.60	61.0	59.8	59.8	62.0	62.37
Carbon, per cent.	5.60	5.16	5.34	5.28	5.67	5.81	5.50	5.55	5.70	5.68	5.53
Slag:											
Chromium, per cent.	15.70	11.65	12.77	7.65	14.84	8.45	15.77	13.5	13.82	17.15	13.13
Carbon charged per 100 lb chrome ore, lb.	14.6	14.6	14.8	14.5	14.7	13.6	12.2	11.8	13.2	12.5	13.3
Weight ferrochrome, lb.	5,909	5,102	5,574	5,291	5,983	7,765	6,877	7,509	6,801	7,925	
Weight chromium, lb.	3,799	3,208	3,560	3,292	3,768	5,020	4,180	4,497	4,061	4,932	
Weight slag, lb.	10,325	8,610	8,505	9,688	10,200	10,333	10,904	12,967	11,567	12,703	
Chromium recovered, per cent.	68.5	66.6	80.2	65.5	72.5	104.5	71.4	69.0	67.0	76.5	74.15
Kw.-hr. per lb. alloy	3.6	4.0	3.6	4.0	3.4	2.7	3.2	3.1	3.4	2.8	3.38

Note.—Recovery is based on metal output, and ore and slag charged. The operations were conducted in one 450-kw. furnace and one 750-kw. furnace.

cent. carbon grade varies from 70 to 80 per cent. Within limits, the grade of ore smelted has no appreciable effect on the recovery when 8-per cent. carbon product is being made.

As shown in Table 3, ore containing from 42.25 to 51.40 per cent. Cr_2O_3 was smelted with no appreciable difference in recovery. With

a drop to 35 per cent. Cr_2O_3, appreciable loss occurs, owing to the greater volume of slag, the latter averaging about the same in chromium as that from the higher grade ore. These same conditions exist when producing ferrochrome, with 6 per cent. carbon as shown in Table 4,

TABLE 5.—*Production of Ferrochrome, with Natural Slag*

	21	22	23	24	25	26	27	28	29	Average
Charge:										
Ore, lb.	14,025	11,050	8,500	14,153	13,655	13,430	12,714	15,300	13,388	
Coke, lb.	4,045	3,268	2,000	3,646	3,514	3,504	3,442	3,600	3,897	
Slag, 12 per cent. Cr, lb.	3,175	1,950	1,500	3,836	3,749	3,999	3,894	2,700	2,363	
Ferrochrome:										
Chromium, per cent.	63.60	58.30	62.1	66.22	67.10	67.20	64.7	61.0	61.20	63.49
Carbon, per cent.	5.50	6.33	5.08	5.60	5.48	5.22	6.73	5.13	5.20	5.58
Slag:										
Chromium, per cent.	13.41	8.50	10.23	7.0	5.4	7.17	7.17	7.90	10.26	8.56
Carbon charged per 100 lb. chrome ore, lb.	16.4	16.9	24.2	14.7	14.6	14.8	15.4	13.4	16.8	16.3
Weight ferrochrome, lb.	5,791	4,390	2,985	5,499	6,150	6,169	5,067	4,533	5,366	
Weight chromium, lb.	3,669	2,553	1,844	3,648	4,103	4,121	3,266	2,870	3,273	
Weight slag, lb.	7,228	.....	.....	7,440	7,027	6,120	4,799	.....	8,382	
Chromium recovered, per cent.	78.0	64.6	69.2	64.5	84.4	89.5	73.3	59.7	74.5	73.0
Kw.-hr. per lb. alloy	3.63	3.42	4.09	4.18	3.58	4.06	4.74	3.32	3.54	3.9

Analyses of Natural Slags

	Per Cent.	Per Cent.	Per Cent.
Cr_2O_3	8.80	13.08	11.91
SiO_2	22.10	23.20	21.68
FeO	7.71	7.71	3.47
Al_2O_3	34.80	35.50	41.20
CaO	5.70	2.30	4.40
MgO	23.18	20.28	20.79

Note.—Recovery is based on metal output, and ore and slag charged. The operations were conducted in three 150-kw. furnaces and three 250-kw. furnaces.

where the ore varied from 34 to 48 per cent. Cr_2O_3. Future operation on nothing but 35-per cent. ore may alter this conclusion.

The addition of lime to the charge unquestionably results in a higher recovery, but increases the carbon content of the alloy. As little as

¼ lb. lime per 100 lb. ore may increase the carbon in the product from 6 per cent. to 7 per cent., charging the same amount of carbon. The recovery, however, is considerably increased, owing to the more fluid

TABLE 6.—*Production of Ferrochrome with Silica Slag*

	30	31	32	33	34	35	36	37	38	Average
Charge: Ore, lb.	51,395	59,330	41,888	40,532	57,902	64,924	43,928	73,814	39,086	
Coke, lb.	13,253	15,213	10,800	10,359	12,514	14,268	9,914	16,552	10,025	
Lime, lb.	1,437	1,618	(*a*)604	1,065	209	1,321	273	81	1,349	
Sand, lb.	4,928	3,945	2,316	2,810	6,202	7,214	4,297	8,088	4,342	
Slag (16% Cr), lb.	9,099	8,670	7,822	7,408	10,188	11,467	7,752	12,728	7,074	
Carbon, lb. per 100 lb. chromite	14.8	14.6	14.7	14.6	12.4	12.6	12.9	12.8	14.6	
Analysis ore: Cr_2O_3, per cent.	42.15	44.95	44.95	42.15	41.74	41.74	43.96	43.96	42.15	
FeO, per cent.	11.27	11.80	11.80	11.27	11.00	11.00	.....	.....	11.27	
Al_2O_3, per cent.		19.70	19.70							
MgO, per cent.		15.50	15.50							
SiO_2, per cent.	8.52	5.80	5.80	8.52	12.46	12.46	9.40	9.40	8.52	
S, per cent.		0.03	0.03							
P, per cent.		0.024	0.024							
Cu, per cent		0.004	0.004							
Ferrochrome Cr, per cent.	62.02	63.82	63.45	62.01	59.78	64.95	58.55	59.50	64.5	62.06
C, per cent.	6.17	5.83	5.73	5.69	5.98	6.36	5.93	5.37	5.80	5.87
Si, per cent.	2.17	1.50	2.07	0.95	1.06	2.83	1.99	1.66	1.48	1.85
S, per cent.	0.09	0.09	0.057	0.041	0.044	0.018	.....	0.014	0.038	0.050
Analysis slag: Cr_2O_3, per cent.	26.63	19.00	20.00	17.52	19.65	18.60	20.00	20.34	21.60	
Al_2O_3, per cent.		38.20	36.00							
FeO, per cent.		8.53	8.74	.....	.....	.....	7.45			
SiO_2, per cent.	15.00	17.50	17.64	21.2	26.16	26.20	27.70	28.62	30.00	
CaO, per cent.	.....	1.00	1.00							
MgO, per cent.	.....	15.00	18.25							
Chromium lost in slag, per cent.	38.25	24.00	24.70	27.00	25.8	25.7	26.8	24.5	32.60	27.71
Chromium recovered, per cent.(*b*)	61.75	76.00	75.30	73.00	74.2	74.3	73.2	75.5	67.4	72.29
Chromium actually tapped, per cent.	71.10	68.37	74.80	70.00	69.1	83.3	64.8	62.8	75.06	71.02

(*a*) Test No. 32 contained also 116 lb. fluorspar. (*b*) Calculated from slag loss.

slag and smaller loss of metallics in the slag. Fluorspar has the same effect as lime.

When operating with a natural slag, *i.e.*, slag produced from the natural constituents of the ore and the ash of the coke, a 6-per cent. carbon product can be made easily, but the loss of metallics is high with most domestic ores, owing to the thick, sticky slag formed. However, some

ores, higher than usual in silica, run very well. Results of operation with a natural slag are shown in Table 5.

On the whole, when producing ferrochrome with 6 per cent. carbon, a small quantity of silica and lime make the most satisfactory flux, if any flux is necessary. The recovery is a little higher than where running with a natural slag, and the presence of silica in slight excess tends to throw carbon and sulfur out of the alloy. Ferrochrome sometimes contains over 2 per cent. silicon when sand is charged, but this is not objectionable. Results of operation with sand are given in Table 6, showing analyses of the slags, the grade of alloy, and the recovery of chromium. The coke contained 57.0 per cent. fixed carbon, 37.9 per cent. ash, 5.27 per cent. volatile matter, and 0.51 per cent. sulfur. The lime analyzed 85 per cent. CaO, and the fluorspar 82 per cent. CaF_2.

The tests given in Tables 3 and 4 were selected at random. When making 8-per cent. carbon product, the slags average 5.21 per cent. chromium as against 13.13 per cent. chromium when producing 6-per cent. carbon material, a difference of 7.92 per cent. The chromium recovery on ferrochrome of 8-per cent. carbon is 94.07 per cent., in comparison with 74.15 per cent. when producing material, containing 6 per cent. carbon, or 19.92 per cent. higher recovery. The effect of a greater amount of carbon in the charge is shown clearly. In Table 3, 17.3 lb. carbon was charged per 100 lb. of chromite, resulting in a recovery of 94.07 per cent. In Table 4, only 13.13 lb. carbon was charged per 100 lb. of chromite, giving a recovery of 74.15 per cent. The ore averaged 44 per cent. chromium.

Ferrochrome containing 6 per cent. carbon can be made easily by using a natural slag. The chromium content of such slag is 8.56 per cent. (Table 5), while, when using lime and silica, the slag contained 13.13 per cent., and with silica alone as a flux (Table 6) it varied from 17.5 to 26.6 per cent. chromic oxide (12 to 18.2 per cent. chromium). In the three cases, the actual chromium tapped was respectively 73.0, 74.15, and 71.02 per cent. The combination of lime and silica thus shows a little higher recovery; generally, however, if the slag will run, it is advisable to use a natural slag. When the alumina is about 35 per cent. and the silica 20 to 25 per cent., these slags are very fluid, but at times retain metallics. The use of a large quantity of silica does not improve the recovery, as shown in Table 6, but it does have a marked effect on the sulfur content of the alloy.

Smelting of Low-grade Ore or Chrome Slag

There has been considerable discussion as to how low-grade chrome ore can be used in the manufacture of ferrochrome. In Table 7, results of smelting chrome slag are given. On slags containing from 13.9 to 36.25 per cent. Cr_2O_3, average 24.37 per cent., an average recovery of 80.2

TABLE 7.—*Smelting Chromium Slag*

	39	40	41	42	43	44	Average
Charge:							
Slag, lb.	17,170	13,770	15,470	14,750	18,000	15,600	
Coke, lb.	2,912	2,340	2,762	1,311	1,472	1,264	
Sand, lb.	1,060	824	930	375	456	398	
Precentage Cr_2O_3 in slag smelted	36.25	31.50	26.10	23.4	13.90	15.10	24.37
Ferrochrome:							
Chromium, per cent.	62.25	61.60	61.60	63.05	60.50	59.05	61.34
Carbon, per cent.	6.55	6.50	6.47	8.12	6.37	6.47	6.81
Slag:							
Chromium, per cent.	9.05	11.70	3.05	3.20	5.70	3.10	5.96
Weight ferrochrome, lb.	5,385	3,270	3,995	3,362	2,031	2,323	
Weight chromium, lb.	3,340	2,002	2,460	2,122	1,228	1,372	
Weight slag, lb.	10,119	8,328	10,577	7,702	8,018	7,563	
Chromium recovered, per cent.	78.6	67.2	88.6	89.4	72.3	85.3	80.2
Kw.-hr. per lb. alloy	2.8	4.0	3.3	3.7	6.4	5.7	4.3

per cent. was made, with the production of metal containing an average of 61.34 per cent. chromium, 6.81 per cent. carbon, 0.063 per cent. sulfur, and 2.5 per cent. silicon. Even with a very low-grade slag, 15 per cent. Cr_2O_3, a recovery of 85 per cent. was made. The cost of actual operation was not much greater than when smelting ore, but including interest and depreciation charges, the total cost would be considerably higher because of the small output of metal for the power consumed.

From these results, it seems reasonable to expect that a good recovery could be made from ore containing 25 per cent. Cr_2O_3, if the product were allowed to contain 7 per cent. carbon.

Grade of Ferrochrome Made from Domestic Ore

Ferrochrome made from domestic chrome ore ranges from 60 to 67 per cent. chromium, as compared with 65 to 70 per cent. chromium in ferrochrome made from imported ore. The lower chromium content is due to the higher percentage of iron in domestic ore, and also probably, to a less degree, to the larger volume of slag, and consequent greater loss of chromium. Aside from the chromium content, the carbon and other impurities are as low as in ferrochrome made from foreign ore. In Table 8, analyses of ferrochrome made from domestic ores are given, the analyses covering material from 1 ton to a carload in quantity.

Use of 60 to 65-Per Cent. Ferrochrome Instead of 65 to 70-Per Cent. in Steel Manufacture

In general, it may be stated that ferrochrome containing 60 to 65 per cent. chromium is not undesirable in the steel industry by reason of

its lower chromium and high carbon content. In the first place, its carbon content need be no higher than that of an alloy containing 65 to 70 per cent. chromium. That is, the carbon in both products can readily be kept under 6 per cent., but, of course, if the standard is maintained at 6 per cent. a larger amount of carbon is present per pound of chromium in the case of the 60 to 65-per cent. ferrochrome. If a user would specify 60 to 65 per cent. ferrochrome and 5.5 instead of 6 per cent. carbon, the

TABLE 8.—*Ferrochrome Made From Domestic Ore*

	Carbon 1 Per Cent. Maximum	Carbon 2 Per Cent. Maximum	Carbon 5 Per Cent. Maximum	Carbon 6 Per Cent. Maximum	Carbon 7 Per Cent. Maximum	Carbon 8 Per Cent. Maximum
Carbon	0.75	1.83	4.75	5.52	7.00	7.67
Chromium	67.20	62.10	63.40	65.00	66.00	66.40
Silicon	0.42	0.40	0.17	0.47	0.93	0.86
Manganese	.....	0.24	0.25	0.29	0.24	0.27
Sulfur	0.052	0.056	0.082	0.078	0.055	0.056
Phosphorus	0.065	0.047	0.045	0.036	0.035	0.036

ratio of chromium to carbon would be the same as in the alloy containing 65 to 70 per cent. chromium and 6 per cent. carbon, and no more carbon would be added to the steel bath by the low-grade than by the high-grade alloy. The carbon content of ferrochrome depends considerably upon the amount of chromium present, which combines with the carbon to form carbides; also upon the amount of silicon present. In the lower-chromium alloy, there should therefore be less carbon than in the high-chromium alloy. It is entirely feasible to make 60 to 65-per cent. ferrochrome with a maximum of 5.5 per cent. carbon. The silicon content, if allowed to increase a little above 1 per cent., also tends to throw the carbon out of the alloy.

Ferrochrome with 67.5 per cent. chromium and 6 per cent. carbon contains 0.088 lb. of carbon per pound of chromium; ferrochrome with 62.5 per cent. chromium and 6 per cent. carbon contains 0.096 lb. of carbon per pound of chromium. The 62.5-per cent. alloy, in order to contain the same amount of carbon per pound of chromium as the 67.5-per cent. alloy with 6 per cent. carbon, should contain 5.5 per cent. carbon. The use of ferrochrome containing 62.5 per cent. chromium and 6 per cent. carbon, in ordinary chrome-vanadium steel, which contains 1 per cent. chromium, will increase the carbon content of the steel above that when the 67.5-per cent. alloy is used, to the extent computed below.

One hundred pounds of 1-per cent. chromium steel will require the addition of approximately 1.48 lb. of 67.5-per cent. ferrochrome; or approximately 1.6 lb. of 62.5-per cent. ferrochrome. The amount of carbon added to 100 lb. of the steel by the use of the 67.5-per cent.

ferrochrome is 1.48 × 6 per cent. = 0.088 lb. The amount of carbon added by the use of the 62.5 per cent. alloy is 1.6 × 6 per cent. = 0.096 lb. The use of the lower grade ferrochrome will thus increase the carbon in a 1-per cent. chromium steel by approximately 0.008 per cent. Or, for example, if a chrome-vanadium steel, made from 67.5-per cent. ferrochrome, contains 0.4 per cent. carbon, 0.15 per cent. vanadium, and 1 per cent. chromium, it would contain, if made from 62.5-per cent. ferrochrome, approximately 0.408 per cent. carbon. No allowance is made for loss of chromium in the addition of the ferrochrome.

It can safely be stated that the use of 60 to 65-per cent. alloy will not increase the carbon in chrome-vanadium by over 0.01 per cent. Also, if

TABLE 9.—*Raw Materials for Making Ferromanganese*

ANTHRACITE	PER CENT.	LIMESTONE	PER CENT.	FLUORSPAR	PER CENT.
Fixed carbon	82.76	CaO	54.56	CaF_2	81.8
Ash	10.20	SiO_2	0.18		
Volatile	7.04	MgO	0.25		
Sulfur	0.84				

TABLE 10.—*Domestic Manganese Ore from Colorado, Utah, and Nevada*

	Homansville, Utah				Las Vegas, Nev.				Montrose, Colo.	
Mn, per cent	39.17	40.62	37.52	37.47	40.37	39.62	38.47	38.75	43.58	44.67
SiO_2, per cent	12.21	11.52	13.02	11.69	13.55	14.16	15.18	15.26	6.58	6.76
Fe, per cent	1.35	1.20	1.30	1.15	0.85	0.75	0.70	.....	4.17	2.79
CaO, per cent	.....	5.78	.....	7.65						
P, per cent	0.285	0.228	0.329	0.351	0.022	0.023	0.023	0.024	0.55	0.145
H_2O, per cent	16.1	14.5	15.5	14.8	8.35	9.15	14.16	14.6		8.30

a 62.5-per cent. alloy containing 5.5 per cent. carbon were used, the steel would contain the same amount of carbon (0.4 per cent.) as if the higher chromium alloy were used.

FERROMANGANESE

The preparation of ferromanganese in the electric furnace consists in charging a proper mixture of ore, coal, and limestone in proper proportions for the reduction of manganese with a minimum slag loss. The furnace operates at 75 to 100 volts. With dioxide ores reduction takes place according to the reaction:

$$MnO_2 + 2C = Mn + 2CO.$$

Iron is added either as a constituent of the ore or in the form of iron turnings. The power consumption varies from 4000 kw.-hr. per long

ton in a 3000-kw. furnace to 7000 kw.-hr. per long ton in a 1000-kw. furnace. The electrode consumption is high, ranging from 150 to 250 lb. per long ton of product, when using amorphous carbon electrodes.

TABLE 11.—*Smelting Domestic Manganese Ore in the Electric Furnace*

	1	2	3	4	5	6	7	8	9	10	Average
Charge:											
Manganese ore, lb.	16,000	18,400	21,600	18,099	17,700	17,700	20,000	21,322	17,500	16,700	
Coal, lb.	4,140	4,600	5,950	4,460	4,400	4,400	4,840	5,280	4,400	3,980	
Limestone, lb.	5,850	6,500	8,450	6,500	7,300	6,500	7,075	7,375	6,000	5,400	
Iron turnings, lb.	1,125	1,250	1,550	900	825	600	660	600	600	540	
Carbon per 100 lb. ore.	25.8	25.0	27.6	24.7	25.8	25.8	24.2	24.7	25.2	23.8	
Manganese in ore, %	39.3	39.9	38.7	39.50	39.5	39.25	39.0	38.6	37.8	37.8	38.93
Ferromanganese											
Manganese, %	76.4	75.6	78.53	79.25	79.46	80.15	80.65	80.5	81.46	81.4	79.34
Silicon, %	1.5	1.1	1.34	0.91	0.47	1.26	0.12	1.29	2.57	1.50	1.20
Phosphorus, %	0.227	0.210	0.173	0.183	0.203	0.205	0.220	0.207	0.198	0.203	0.202
Sulfur, %	.....	0.005	0.015	0.030	0.013	0.017	0.004	0.017	0.035	0.024	0.018
Slag:											
Manganese, %	12.58	15.38	9.08	12.88	16.18	13.78	14.97	11.56	12.05	10.95	12.94
SiO_2, %	28.84	27.36	28.74	27.90	25.64	27.54	22.74	28.96	29.52	30.72	27.79
CaO, %	38.06	33.22	30.51	35.03	35.48	37.25	33.95	37.49	35.96	37.09	35.40
MgO, %	.....	.....	9.02	.....	.....	2.60	.....	.....	.....	.....	5.81
Al_2O_3, %	.....	.....	8.90	.....	.....	9.16	.....	.....	.....	.....	9.03
Wt. ferromanganese, lb.	6,785	6,552	7,674	6,502	7,823	6,089	6,937	7,564	5,420	6,865	6,821
Weight slag, lb.	10,556	9,221	11,043	8,537	10,903	8,190	10,341	9,145	8,353	9,753	9,604
Manganese lost in slag, %	20.7	19.3	11.9	15.5	25.2	16.2	19.8	12.8	15.5	16.8	17.3
Manganese lost by volatilization and dust, %	.....	10.2	16.9	12.1	.....	13.8	8.7	13.6	17.3	.....	7.3
Manganese recovery, %	82.3	70.5	72.1	72.4	88.4	70.0	71.5	73.6	67.2	86.5	75.4
Kw.-hr. per long ton ferromanganese	6,400	7,150	5,980	6,220	6,320	6,600	6,420	6,400	6,450	7,350	6,429

Analyses of some domestic manganese ores produced in Colorado, Utah, and Nevada are given in Table 10. Some of this ore is so high in phosphorus that it has to be mixed with the low-phosphorus ore to yield a product under 0.2 per cent. phosphorus. Most of these ores are very dusty, resulting in a [illegible] dust loss, even in the electric furnace

Smelting Domestic Manganese Ore in the Electric Furnace

Table 11 gives results from the electric smelting of the ores given in Table 10, and the other materials shown in Table 9. These results are based on the first month's operation of a 1200-kw. furnace, and do not represent what is being done by the same furnace at the present time, when a considerably better recovery is being made with a lower power consumption.

Smelting an ore averaging 38.9 per cent. manganese and 13.0 per cent. SiO_2, the recovery in metal tapped is 75.4 per cent., with an average loss of 17.3 per cent. in the slag and 7.3 per cent. mechanically and by volatilization. Considerably better results are now being obtained, especially as regards slag loss. The power consumption was 6429 kw.-hr. per long ton of ferromanganese, measured on the primary side of the transformers. The average metal produced contained: manganese, 79.34; silicon, 1.2; phosphorus, 0.202; sulfur, 0.018 per cent. The quantity of coal charged was less than 10 per cent. above the theoretical amount required for complete reduction, which is contrary to the practice at several electric-furnace plants.

Ferromolybdenum

The use of ferromolybdenum in the metallurgy of steel is in its infancy. Previous to 1914, probably not over 10 tons was produced yearly in the United States, practically all of this being exported; but with the high price of tungsten, search for molybdenum ores became more active, and now several hundred tons per year are made in this country, practically all of which is exported. The alloy has not been widely used because of the supposed scarcity of ore, and the prejudices of American steel manufacturers against it, owing to difficulties encountered in its use in steel. The scarcity of ore has been overcome, and at present Colorado is the largest producer of molybdenum in the world. The prejudices of American steel manufacturers have not, however, been overcome.

Ferromolybdenum was made from roasted molybdenite in the crucible before the introduction of the electric furnace, but can now be made directly from raw sulfide in the electric furnace. Another source of molybdenum is lead molybdate, wulfenite, which is fused with soda ash and carbon to produce lead bullion and sodium molybdate slag. The slag is then smelted in the electric furnace, with carbon as a reducing agent, and suitable fluxes, to produce ferromolybdenum.

The standard grade of ferromolybdenum is not well established. Some manufacturers make a product containing 50 to 80 per cent. molybdenum and 3 per cent. carbon and manage to sell it. Others, by different methods, make a product containing less than 1 per cent. carbon. A great deal of ferromolybdenum containing 0.25 per cent. sul-

fur has been sold but most consumers will not accept a product containing over 0.1 per cent. sulfur. Ferromolybdenum containing 80 per cent. molybdenum has a dull gray iron color, coarse structure, high density, and is non-magnetic. It does not break easily.

Production of Ferromolybdenum in the Electric Furnace

Raw materials used in preparation of ferromolybdenum may be: for ore, the sulfide, molybdenite, or sodium molybdate slag made from wulfenite; for reducing agent, some form of carbon, or 90-per cent. silicon metal ground to pass 60 mesh; for fluxes, lime and fluorspar. There are two methods of manufacture:

1. Reduction with carbon and excess of lime, according to the reaction:

$$2MoS_2 + 2CaO + 3C = 2Mo + 2CaS + 2CO + CS_2$$

Pure molybdenite contains 60 per cent. molybdenum and 40 per cent. sulfur. According to the above reaction, 100 parts of molybdenum are reduced from 170 parts of molybdenite by 18.8 parts of carbon. For every 100 parts of molybdenum, 58 parts of lime are necessary for slagging the sulfur as calcium sulfide. This reaction works out closely to the theoretical, and there is no difficulty in making a product with about 0.1 per cent. sulfur: the product will contain 1.3 to 3 per cent. carbon. If a lower-carbon alloy is desired, the crude metal is broken up and refined with an oxidizing slag of iron ore. The amount of iron in the alloy is varied as desired by the addition of iron turnings in the smelting furnace, or of iron oxide in the refining furnace.

2. Reduction with silicon metal, according to the reaction:

$$MoS_2 + Si = Mo + SiS_2.$$

This method has been used recently in the production of 50-per cent. ferromolybdenum, although when this grade is being made, ferrosilicon may be used. Lime is sometimes added to help slag the sulfur as calcium sulfide. By this method, the production of 100 parts of molybdenum requires 29 parts of silicon.

When sodium molybdate slag is the source of molybdenum, reduction takes place according to the reaction:

$$Na_2MoO_4 + 3C = Mo + 3CO + Na_2O.$$

The use of sodium molybdate requires considerably more power than the other reactions, because very little reduction occurs until all free soda salts which may be in the slag are driven off. The regular grade of sodium molybdate slag used for this purpose contains 30 to 40 per cent. MoO_3. In all electric-furnace work, the presence of sodium salts interferes with the speed of the reactions.

The average wulfenite[1] from which ferromolybdenum is produced

[1] Alan Kissock, Manager, Molybdenum Products Co., Tucson, Arizona; personal correspondence.

contains: MoO_3, 16; lead, 50; SiO_2, 6; FeO, 11.0; CaO, 2.0; arsenic, 0.8; phosphorus, 0.05 per cent. This concentrate is smelted in a circular, water-jacketed lead blast furnace with coke and soda ash to produce lead bullion and molybdenum slag. The slag contains: MoO_3, 33; lead, 1.0; SiO_2, 11 to 14; FeO, 17; CaO, 7; arsenic, 1.0; phosphorus, 0.1 per cent.; remainder, soda. This molybdenum slag, crushed to about ½ in., is smelted in single-phase electric furnaces of the Siemens type, lined with magnesite, to produce ferromolybdenum. The charge is calculated to a monosilicate slag, using lime as a flux. Iron ore is added to produce an alloy containing 60 to 65 per cent. molybdenum, and residues from an oil-gas plant are used for the reducing agent.

The average power consumption is 7 to 7.5 kw.-hr. per pound of molybdenum produced. The recovery varies from 78 to 80 per cent., with a loss of 10 per cent. in slag, and about 10 per cent. mechanically and by volatilization.

TABLE 12.—*Ferromolybdenum Made from Sodium Molybdate Slag*

Molybdenum, per cent	61.0	62.34	66.96	64.78	1.91	62.92
Carbon, per cent	2.31	2.214	2.03	1.51	62.35	1.29
Silicon, per cent	0.41					
Phosphorus, per cent	0.086	0.018	0.029	0.018	0.040	0.13
Sulfur, per cent	0.098	0.075	0.060	0.022	0.049	0.030
Arsenic, per cent	1.10					
Copper, per cent	0.018					
Manganese, per cent	0.2					

Ferromolybdenum containing 80 per cent. molybdenum and under 1 per cent. carbon cannot be regularly tapped from the electric furnace because of its high melting point, so that when this grade of alloy is to be made, the furnace must be of the knock-down variety, for removal of the button. The slag is tapped off, and when this operation is finished the metal is dug out. A 50 to 60-per cent., low-carbon product can be tapped, and a considerable quantity of this grade is made in tapping furnaces.

Molybdenum in Steel Manufacture

Ferromolybdenum is added to steel as a fixed addition, nearly all of the molybdenum remaining in the steel. It is supposed to give the steel properties similar to those of tungsten steel, but only one-third to one-half as much molybdenum is necessary; that is, where regular high-speed steel contains 18 per cent. tungsten, 6 to 9 per cent. of molybdenum may be substituted. However, it gives these properties only when the addition is properly made and proper heat treatment follows. The regulation of these factors caused so much trouble and expense that, in this country, the manufacture of molybdenum high-speed tool steels

has been practically discontinued for several years. It is used for this purpose abroad, however, to a considerable extent. At the present time, it is mainly employed in tool steel as an auxiliary rather than as a major constituent.

Various reasons have been assigned for the discontinuance of the use of molybdenum in these steels. Taylor found that molybdenum in rapid steels caused irregular performance; that steels of the same composition and having had seemingly the same treatment gave large variations in their maximum cutting speeds. One manufacturer has stated that the ingots crack in forging, the tools crack on quenching, and molybdenum appears to volatilize from the steel when heated; the latter might be due to the production of molybdenum oxide, which is much more volatile than the metal itself.

When small quantities of molybdenum, say 0.25 per cent., are added, the elongation and elastic limit of the steel are greatly increased. When molybdenum is combined with nickel, the resistance to shock is increased without diminishing the elongation. Its utilization for linings of big guns was originated by the Germans, with such success that the Allies are said to use it now for the same purpose. This may account for the heavy exports of molybdenum.

Ferrotungsten

Tungsten is the leading alloy metal used in the manufacture of high-speed steel. Probably about half of the consumption is used in the form of ferrotungsten, the remainder being tungsten metal. Colorado has supplied a large part of the tungsten used in steel manufacture, but is gradually being surpassed by California, where the large scheelite deposits can produce tungsten considerably cheaper. Ferberite, found in Boulder County, Colo., is one of the most favorable forms of tungsten for production of ferrotungsten, containing both iron and tungsten in the proper proportion to yield 70 to 80 per cent. ferrotungsten. At present, three electric-furnace plants in Colorado are making or have made ferrotungsten. The standard grade produced by the Ferro Alloy Co. contains 75 per cent. tungsten, 0.8 carbon, 0.4 silicon, 0.5 manganese, 0.01 sulfur, 0.02 per cent. phosphorus. The Tungsten Products Co. makes about the same grade of product, except that the carbon is kept below 0.5 per cent., and the tungsten runs as high as 85 per cent. The product of this company is probably the highest grade and the purest ferrotungsten made in this country. Ferrotungsten has a high density, fine gray fracture, and is not crystalline—the higher the carbon, the coarser the fracture. The carbon in ferrotungsten occurs as the double carbide, $Fe_3C.W_2C$.

One method commonly employed for production of ferrotungsten is reduction of [redacted] or some other concentrate in the electric furnace,

with carbon as a reducing agent, followed by a subsequent refining and decarburization. Reduction with carbon takes place according to the following reaction:

$$Fe_2WO_4 + 4C = Fe_2W + 4CO.$$

Theoretically, the product would contain 62.3 per cent. tungsten, but, owing to the variation in iron in the ferberite, and to the fact that tungsten reduces from ferberite, in the electric furnace, before the iron comes down, a considerable part of the iron passes into the slag, so that the alloy contains about 70 per cent. tungsten. The slag may contain as high as 8 per cent. FeO, but less than 1 per cent. WO_3. Theoretically, the reduction of 100 parts of alloy from 122 parts of ferberite requires 16 parts of carbon, but practically about 25 per cent. excess of carbon is charged. Small amounts of lime and fluorspar are used to flux the silica. Operating in this manner, with an excess of carbon, a product is made containing 3 per cent. carbon, 70 per cent. tungsten, 0.05 per cent. phosphorus, and 0.01 per cent. sulfur: the slag contains below 1 per cent. WO_3. If desired, this product can be tapped from the furnace, but, as its behavior is erratic, it is preferable to allow the metal to collect in the bottom of the furnace, forming, in a 150-kw. furnace, a button about 3 ft. in diameter, 6 in. thick, and weighing 1000 to 1200 lb.; the slag is poured off by tilting as it fills the furnace.

The concentrate used in the runs given in Tables 13 and 14 contained WO_3, 60.36 per cent.; iron, 22.0; SiO_2, 8.0; manganese, 0.5; sulfur, 0.35; phosphorus, 0.05 per cent. The analysis of the coke was: fixed carbon, 81.8; volatile, 1.70; ash, 16.5; sulfur, 0.60; phosphorus, 0.092 per cent. The lime contained: CaO, 89.0; MgO, 0.73; SiO_2, 1.1; sulfur, 0.41; phosphorus, 0.072 per cent. The fluorspar contained 80.0 per cent. CaF_2.

Smelting of Ferberite Concentrate

A typical operation is conducted as follows: The initial charge is 65 lb. of a mixture composed of 200 lb. concentrate, 42 lb. coke, 56 lb. lime, 6 lb. fluorspar. Three more 65-lb. charges are added at intervals of ½ hr., and at 2½ hr. from the start, the furnace is tilted and the slag poured. This cycle is repeated until a 1200-lb. button has been formed, requiring 24 to 36 hr. The furnace is allowed to cool, is torn down, and the button of metal removed. This is then cleaned and broken up, the breaking process being somewhat difficult with a 3-per cent. carbon alloy.

The results of smelting ferberite concentrate to produce a crude ferrotungsten are given in Table 13. In this operation there is practically no loss of tungsten in the slag, the slags averaging 0.72 per cent. tungsten. Considerable mechanical loss is shown in these runs, but this occurs mainly in dust, when cleaning and breaking the button, and is

recovered by melting. The total loss of tungsten in the smelting operation does not exceed 5 per cent.

TABLE 13.—*Smelting of Ferberite Concentrate*

	1	2	3	4	5	6	7	8
Charge:								
Concentrate, lb.	1,200	1,800	1,600	1,600	2,275	3,150	2,950	3,100
Coke, lb.	252	380	330	330	476	662	662	648
Fluorspar, lb.	96	180	128	130	62	96	90	120
Lime, lb.	336	500	436	436	628	872	810	855
Ferrotungsten:								
Tungsten, per cent.	71.71	71.35 68.3	67.9	68.2				
Carbon, per cent.	1.01	2.6 1.7	1.03	1.09				
Sulfur, per cent.	0.078	0.10	0.048	0.068				
Phosphorus, per cent.	0.081	0.112	0.065	0.11				
Slag:								
Tungsten, per cent.	0.56	0.80	0.90	1.26	0.20	0.88	0.35	0.39
Weight ferrotungsten, in button, lb.	731	982	847	908	954	1,884	1,490	1,971
Weight ferrotungsten, poured, lb.		122						
Weight, slag, lb.	588	1,020	510	1,121	1,960	1,612	1,200	2,613
Length run, hr.	19.33	26.17	20.17	22.0	36.0	40.5	44.0	48.0
Power on, hr.	17.33	21.08	18.0	19.42	34.83	38.83	42.5	46.33
Tungsten lost in slag, per cent.	0.57	0.95	0.63	1.96	0.36	0.94	0.29	0.66
Tungsten lost mechanically, per cent. (*a*)	10.23	11.65	19.37	11.84				
Tungsten recovered, per cent.	89.2	87.4	80.0	86.2				

(*a*) Most of this is recovered in dust.

Note.—The current consumption was 130 kw. at 95 volts.

Refining Ferrotungsten

The crude metal, broken to about 6-in. size, is refined as follows: A charge of 150 lb. metal and 75 lb. ferberite concentrate is smelted for ½ hr., when 12 lb. fluorspar is added. After another 3 hr. the slag is poured, and a fresh charge is started. The process is continued for from 36 to 48 hr. until a button weighing 1500 lb. has formed in the furnace, a larger button being permissible because the low-carbon alloy is easier to break. The furnace is allowed to cool, is torn down, and the button of metal removed, cleaned, and broken. The refined button forms very compactly, and is free from slag. The refining has reduced the carbon from 3 to 0.8 per cent., reduced the phosphorus from 0.05 to 0.01 per cent., left the sulfur the same, and increased the tungsten from 70 to 75 per cent. The refining slag contains 5 to 20 per cent. tungsten, and is [redacted] in a special run.

Table 14 gives data on the refining of crude ferrotungsten containing 2 to 3 per cent. carbon, to yield a product with less than 1 per cent. carbon.

TABLE 14.—*Refining Ferrotungsten*

	15	16	17	18	19	20	21	22	2	24	25	26	27	28
Charge:														
Ferrotungsten, lb.	1,400	1,200	800	1,835	1,680	1,160	1,190	1,050	1,200	1,200	1,200	1,200	1,350	1,500
Tungsten concentrate, lb.	448	480	320	720	674	462	476	420	520	670	600	675	750	825
Fluorspar, lb.	140	120	80	180	141	96	96	84	96	108	96	1 8	108	120
Ferrotungsten:														
Carbon, in button, per cent.	0.84	0.88	0.90	0.68	0.71	0.73	0.88	0.75	0.65	0.66	0.91	0.66	0.97	0.57
Carbon, in poured, per cent.	1.37	0.95	1.27	1.32	1.22	.....	.....	.....	.....	1.02	1.59	1.50	.....	0.63
Slag:														
Tungsten, per cent.	5.21	5.21	5.21	5.21	9.50	17.18	.....	5.54	5.54	.....	.....	.....	.....	.....
Concentrate charged per 100 lb. ferrotungsten.	32.00	40.00	40.00	39.25	40.25	39.75	40.20	40.0	43.4	55.8	50.0	56.3	55.6	55.0
Weight ferrotungsten in button, lb.	939	1,232	400	1,868	1,424	814	727	913	922	1,217	1,278	1,234	1,535	1,672
Weight ferrotungsten poured, lb.	.....	.....	720	186	335	345	70	.....	138	101	62	.....	266	107
Weight slag, lb.	393	497	534	812	568	544	412	373	495	543	418	576	470	571
Length run, hr.	22.66	21.33	19.5	17.33	34.83	33.75	26.0	33.5	29.5	34.5	35.33	32.33	33.75	37.33
Power on, hr.	22.00	20.83	19.0	16.17	33.83	32.66	25.45	31.83	28.75	33.33	34.5	31.5	33.0	36.5

Note.—The power consumption was 130 kw. at 95 volts.

During these runs a considerable quantity of ferrotungsten was poured from the furnace with the slag. While the metal obtained in button form contained less than 0.9 per cent. carbon, the poured metal averaged

1.2 per cent. carbon. All of the slag from the refining process, averaging about 10 per cent. tungsten, is re-treated for ferrotungsten, so that there is no loss in refining, except a slag loss of less than 1 per cent., and a mechanical loss. By smelting this slag, ferrotungsten was produced which contained tungsten, 79.1; carbon, 2.40; sulfur, 0.01; phosphorus, 0.02 per cent.; the slag from this operation contained 0.95 per cent. tungsten.

Ferrotungsten is one of the easiest ferro-alloys to manufacture, except for the fact that it must be made in a knock-down furnace, due to the high melting point of the alloy, about 2500° C. Metallurgically it is simple, as tungsten has not so great a tendency to form carbides or to oxidize as have chromium and uranium. Ferrotungsten containing less then 1 per cent. carbon can be made in a single smelting operation, by careful regulation of the carbon in the charge, and the use of an acid slag. The disadvantage of this method is that the product is less pure, and the slag loss is considerably higher; the metal will contain much higher phosphorus and sulfur, and the total loss of tungsten may be 25 per cent. as compared with 10 to 15 per cent. by the two-stage process.

Nearly every manufacturer of ferrotungsten has tried to tap it from the furnace, but none has continued the practice. If the tungsten is reduced to 55 to 60 per cent., the alloy can be tapped, even with carbon as low as 0.2 per cent. With tungsten at 70 per cent., metal containing 0.9 per cent. carbon has run out while pouring slag, but generally any metal which comes out with the slag will contain 1.25 per cent. carbon. The irregularity of tapping operations with ferrotungsten has prevented the adoption of that method, because if part of the metal must be retained in the furnace, it is cheaper to keep it all there. Furthermore, the poured metal is very hard to break.

In ferrotungsten production it is essential to maintain regularity of charging, as this is the only way to insure homogeneous composition of the metal, since the button is not melted throughout. Even during refining, the metal does not melt to any extent, the action taking place while the metal is in a pasty condition. To insure a more homogeneous product, consumers require the alloy to be crushed to ¼ in., this crushing forming one of the main trials of the manufacture.

Analyses of ferrotungsten made from ferberite by smelting and refining are shown in Table 15.

TABLE 15.—*Ferrotungsten Produced from Ferberite*

	Per Cent.	Per Cent.	Per Cent.	Per Cent.	Per Cent.
Tungsten	74.19	71.19	71.50	69.70	72.09
Carbon	1.00	0.99	0.88	0.99	0.96
Silicon	0.39	0.42	0.70	0.44	0.76
Manganese	0.53	0.14	0.21	0.15	0.24
Sulfur	0.010	0.012	0.039	0.002	0.055
Phosphorus	0.013	0.021	0.022	0.024	0.037

Ferrovanadium

All vanadium ores require a chemical treatment to put the vanadium in the form of vanadium oxide or iron vanadate before it can be reduced to metal. The process for extraction of vanadium from Colorado roscoelite[2] has been described as follows:

> Ore containing 1 to 3 per cent. V_2O_5 is treated. The ore is given a roast with salt and iron sulfide, leaving the vanadium in the form of sodium vanadate, which is soluble in water. The iron vanadate is added to aid in keeping a high temperature in the roaster. The roasted ore is leached with a weak solution of sodium vanadate followed by hot water. Before precipitation with iron sulfate, the solution is cooled with an air jet. The iron sulfate must be present in excess, and 4 to 6 hr. are necessary for complete precipitation. Agitation with air continues for 4 to 6 hr. after precipitation to prevent precipitation of lime. The precipitate of iron vanadate is filtered in a Kelley press and washed with water. After drying, it is shipped to the alloy reduction plant.

Probably 75 per cent. of the ferrovanadium produced is made in the open-hearth or crucible furnace, by a modification of the thermit process, using aluminum as the reducing agent. The remainder is produced by the electric furnace, using 90 per cent. silicon metal as the reducing agent.

Standard ferrovanadium contains from 30 to 40 per cent. vanadium, and less than 0.5 per cent. carbon, 1 per cent. silicon, 2 per cent. aluminum, 0.1 per cent. sulfur, and 0.1 per cent. phosphorus. It has a fine fracture, is not crystalline, and is bright gray in color. Experience has shown that the 33-per cent. grade is best for making vanadium additions to steel. Carbon is very objectionable in ferrovanadium, because over 1 per cent. carbon causes some of the vanadium to enter the steel in the form of carbide, producing an injurious effect.

The largest output of ferrovanadium is made by reduction of vanadium oxide or iron vanadate with aluminum shot in a gas-fired open hearth, slagging off the alumina by addition of soda ash or fluorspar. If vanadium oxide is used, iron turnings are added as necessary. The process is partly exothermic, but not enough heat is produced by the reaction between the aluminum and the vanadium oxide to carry on the reduction without the application of heat The use of iron vanadate is open to the objection that additional aluminum is required for the reduction of iron, which increases operating costs considerably at the present time. Reduction takes place according to the reaction:

$$3V_2O_5 + 10Al = 6V + 5Al_2O_3$$

In the electric-furnace process, the use of carbon as reducing agent for vanadium oxide has been generally discontinued because of the difficulty in keeping the carbon low, and the tendency to form carbides

[2] *Mining World* (July 17, 1915) **43**, 105.

which the steel manufacturers maintain are injurious to steel. Silicon has been found the best reducing agent, aluminum volatilizing too easily and being more expensive. The raw materials for the manufacture of ferrovanadium in the electric furnace are steel turnings, vanadium oxide, silicon, lime, and fluorspar. The turnings should be low in carbon. While 90 per cent. silicon metal, ground to 60 mesh, is preferable, ferrosilicon can be used if necessary. All of these materials should be low in phosphorus. Any standard type of electric furnace can be used for making ferrovanadium, but it should be lined with magnesite.

TABLE 16.—*Electric Smelting of Vanadium Oxide*

	1	2	3	4	5	Average
Charge:						
Vanadium oxide, lb	396	330	462	528	528	
Steel turnings, lb	190	200	280	320	350	
Silicon metal, lb	252	210	294	333	295	
Lime, lb	780	670	910	980	1000	
Ferrovanadium:						
Vanadium, per cent	29.0	31.4	34.5	32.2	33.80	32.1
Silicon, per cent	3.47	3.01	3.44	6.09	4.09	4.02
Slag:						
Vanadium, per cent	0.36	0.87	0.90	2.91	0.87	1.18
V_2O_5 in vanadium oxide, per cent	83.5	83.5	84.4	84.0	83.3	83.7
Silicon charged per 100 lb. vanadium oxide, lb	63.7	63.7	63.7	63.5	56.0	
Weight ferrovanadium, lb	550.5	344.0	535.0	505.0	504.5	
Vanadium recovered, per cent	86.0	70.1	83.9	66.3	70.0	75.2
Kw.-hr. per lb. ferrovanadium	1.7	2.9	1.9	2.0	2.0	2.1

Electric Smelting of Vanadium Oxide

Steel turnings are first charged into the furnace; when they are melted, vanadium oxide is added, followed by a mixture of silicon and lime. When the slag is fluid, in from 1 to 1½ hr., it is raked off, and more vanadium oxide charged, followed by a mixture of silicon and lime. About an hour after this the slag is pulled out and the metal tapped. Instead of pulling the slag out through a door, a separate slag tap hole may be provided. A large quantity of slag is formed in the operation because of the large quantity of lime which must be charged to combine h the silica formed by reduction of vanadium oxide with silicon. If iron vanadate is used instead of vanadium oxide, no turnings are charged, and the quantity of silicon and lime are increased. This operation is based upon the following reactions:

$$2V_2O_5 + 5Si = 4V + 5SiO_2$$

$$5SiO_2 + [illegible] = [illegible]SiO_3.$$

The ferrovanadium produced in this manner contains from 4 to 8 per cent. silicon, 30 per cent. vanadium, 0.5 per cent. carbon, 0.05 per cent.

TABLE 17.—*Refining Ferrovanadium to Remove Silicon*

	6	7	8	9	10	11	12	13	14	15	Average
Charge:											
Ferrovanadium, lb.	50.00	50.00	50.00	50.00	50.00	50.00	50.00	50.00	50.00	50.00	
Vanadium oxide, lb.	10.00	5.00	5.00	5.00	5.00	5.00	5.00	5.00	7.50	7.50	
Lime, lb.	15.00	7.50	7.50	7.50	7.50	7.50	7.50	7.50	11.50	11.50	
Ferrovanadium:											
Vanadium, per cent	38.50	39.80	39.00	39.65	39.65	35.10	38.75	36.90	36.30	36.40	37.00
Carbon, per cent	0.52	0.88	0.83	0.85	0.85	0.88	0.71	0.54	0.73	1.34	0.81
Silicon, per cent	0.23	0.63	0.89	0.96	0.95	1.01	1.08	0.54	0.59	0.19	0.70
Length run, min.	145.00	115.00	150.00	120.00	104.00	100.00	100.00	150.00	143.0	151.00	
Weight metal, lb.	34.00	49.00	54.00	50.50	47.00	37.50	37.50	41.50	41.50	41.00	
Power consumption, kw.-hr. per lb. ferrovanadium	2.40	1.50	1.60	1.30	1.30	1.50	1.50	2.30	2.00	2.00	1.70
Percentage vanadium charged as ferrovanadium recovered as ferrovanadium	73.10	109.00	117.60	111.80	104.00	73.60	81.10	84.50	84.10	82.50	92.10

sulfur, 0.05 per cent. phosphorus. A product containing 2 to 4 per cent. silicon can be obtained in a single operation by reducing the amount of

silicon charged, and allowing a greater loss of vanadium in the slags. The slags from normal operation contain less than 1 per cent. vanadium.

Table 16 shows an average recovery of 75.2 per cent. of the vanadium, with a power consumption of 2.1 kw.-hr. per pound of ferrovanadium. Smelting of iron vanadate shows similar results, except that there is a considerably greater consumption of silicon metal, owing to the necessary reduction of iron oxide.

Refining Ferrovanadium

To obtain a ferrovanadium containing less than 1 per cent. silicon, the 4 to 8-per cent. product from the smelting furnace is refined in the electric furnace with a slag of vanadium oxide, lime, and fluorspar, giving a product containing 35 per cent. vanadium, 0.9 per cent. silicon, 0.5 per cent. carbon, 0.05 per cent. sulfur and 0.05 per cent. phosphorus. In the refining process, the ferrovanadium is simply mixed with the slag-forming materials in proper proportions, melted in the furnace, and tapped. The removal of silicon in this manner is a simple matter, and much less difficult than refining to remove carbon. The silicon in the ferrovanadium reduces the vanadium from the oxide, and passes off as calcium silicate in the presence of lime. The usual grade of vanadium oxide contains over 80 per cent. V_2O_5. Iron vanadate contains from 25 to 40 per cent. V_2O_5. Slag produced by refining contains about 5 per cent. V_2O_5, and is run back through the refining furnace so that there is no loss of vanadium by this method. Also, there is no danger of reducing the high silica in this slag as it cannot be reduced when silicon is the only reducing agent present.

Results of refining high-silicon ferrovanadium are indicated in Table 17. The ferrovanadium to be refined contained 35.8 per cent. vanadium and 4.95 per cent. silicon; the vanadium oxide contained 75.0 per cent. V_2O_5, and the lime contained 61.45 per cent. CaO. The carbon in the ferrovanadium product was introduced by the electrodes during both smelting and refining, and some of it came from the silicon metal used in smelting, which contained 90 per cent. silicon, and 0.52 per cent. carbon.

The power consumption was 1.7 kw.-hr. per pound of ferrovanadium produced, which, added to 2.1 kw.-hr. for smelting, gives a total of 3.8 kw.-hr. per pound of refined ferrovanadium. The increase in grade of the ferrovanadium from 35.8 to 37.0 per cent. is due to the replacement of silicon by vanadium during refining. Typical slags from the refining of ferrovanadium to remove silicon are as follows:

	1 Per Cent.	2 Per Cent.
V_2O_5	1.44	5.28
SiO_2	28.03	31.37
FeO	2.52	2.19
CaO	53.00	40.90
MgO	.....	12.7

The difficulties encountered in the removal of carbon from ferrovanadium are illustrated in Table 18. The carbon of the ferrovanadium could be reduced from 1.5 per cent. to less than 1 per cent., but a considerable quantity of the vanadium originally present in the alloy oxidized and went into the slag. Also, it was difficult to get clean taps, as the metal did not separate readily from the slag.

TABLE 18.—*Refining Ferrovanadium to Remove Carbon*

	16	17	18	19	22	23	24	25
Charge:								
Ferrovanadium, lb	25.00	25.00	25.00	25.00	25.00	25.00	25.00	25.00
Lime, lb	.....	.....	.....	.....	3.00	3.00	3.00	3.00
Iron scale, lb	1.00	2.00	2.00	2.00	1.50	1.50	1.00	1.00
Vanadium oxide, lb	0.50	1.00	1.00	1.00	1.50	1.50	1.00	1.00
Fluorspar, lb	0.40	0.75	0.75	0.75	0.75	0.75	0.75	0.75
Silica, lb	.....	.....	.....	.....	0.75	0.75	0.75	1.25
Per cent. carbon in ferrovanadium refined	1.45	1.45	1.45	1.45	1.45	1.34	1.34	1.15
Ferrovanadium produced:								
Carbon, per cent	1.09	1.02	1.03	0.97	1.04	0.79	0.83	0.80
Vanadium, per cent	.....	27.15	.....	.....	.....	27.15	24.20	27.15
Silicon, per cent	.....	0.49	.....	.....	.....	0.24	0.36	0.38
Slag:								
SiO_2, per cent	.....	.....	.....	.....	.....	9.16	11.82	27.36
V_2O_5, per cent	.....	.....	.....	.....	.....	22.77	28.29	6.00
FeO, per cent	.....	.....	.....	.....	.....	9.79	8.47	9.79
CaO, per cent	.....	.....	.....	.....	.....	17.48	12.85	9.53
MgO, per cent	.....	.....	.....	.....	.....	16.24	12.80	16.60
Al_2O_3, per cent	.....	.....	.....	.....	.....	2.68	4.34	3.38
Length of run, hr	1.25	3.00	2.50	2.00	2.50	2.25	2.00	2.00
Weight metal, lb	15.00	16.00	21.00	18.00	19.00	21.00	.....	15.00

FERRO-URANIUM

Ore Treatment

All ferro-uranium manufactured in this country is derived from the carnotite ore of Colorado. To put the uranium in a suitable form for the production of ferro-uranium, the ore is first subjected to a wet treatment which has as its main object the extraction of radium, with uranium oxide and vanadium oxide or iron vanadate as byproducts.

In the method of extraction used by the National Radium Institute,[3]

[3] C. L. Parsons, R. B. Moore, S. C. Lind, O. C. Schaefer: Extraction and Recovery of Radium, Uranium and Vanadium from Carnotite. U. S. Bureau of Mines *Bull.* 104 (1915).

a high recovery of radium is the object rather than of uranium and vanadium. The ore is ground to 20 mesh, and is leached with strong hot nitric acid in earthernware pots, the acid being kept near the boiling point with live steam. The charge is stirred with wooden paddles, for about 15 min., and then dumped on an earthenware filter, where it is washed with hot water. The whole operation of leaching, filtering, and washing takes about 7 hr.; a new device for filtering has been installed recently, which reduces the time considerably. The residue goes to the dump, and the solution to wooden tanks, where it is diluted with water. The solution is stirred, and sodium hydroxide runs in slowly to reach as nearly as possible the neutral point without forming a permanent precipitate. Barium chloride and sulfuric acid are stirred in for 1 hr., when the whole solution, containing the barium sulfate precipitate, is elevated to a conical settling tank, where it settles for 4 days.

The solution is decanted into a tank containing boiling sodium carbonate, where the iron, calcium, and most of the aluminum are precipitated, while the uranium and vanadium go into solution as the double carbonate of uranium and sodium, and sodium vanadate. The solution is boiled for 3 hr. after the acid solution is run in.

The radium-barium sulfates and the associated liquor from the conical tank are run onto an earthenware suction filter, where they are filtered, washed, and finally treated with dilute solution of sodium hydroxide to remove the last traces of free acid. The filtrate is run into the carbonate tank, and the sulfates are dried.

The carbonate solution, carrying the uranium and vanadium, is nearly neutralized with nitric acid, the solution being constantly stirred with compressed air. Then sodium hydroxide is added to the boiling solution until precipitation of sodium uranate is complete. The hot solution from the sodium uranate is completely neutralized with nitric acid, air being blown into the liquid in order to eliminate the carbon dioxide. Ferrous sulfate is then added, the liquid being continually agitated, and the iron vanadate precipitate is filtered and washed. The filtrate from the iron vanadate is almost wholly a solution of sodium nitrate, the main impurity being sodium sulfate. The solution is evaporated in iron tanks, and the crystals are used to make fresh nitric acid for use in the plant. The recovery of radium averages 90 per cent. or better; of uranium, about 85 per cent.; and of vanadium, 30 per cent.

Sodium uranate is converted to uranium oxide by fusion with an alkali chloride, such as sodium chloride, and a carbonaceous material which acts as a reducing agent. This is followed by cooling, dissolving the soluble matter, and washing the residue successively with an alkali solution, water, and acid. This residue is high-grade black oxide of uranium, UO_2, which is the concentrated material used in the electric-furnace manufacture of ferro-uranium.

Experimental Development of Ferro-uranium

Ferro-uranium was developed not because of any particular need for uranium in steel manufacture, but because of the large quantity of sodium uranate, $Na_2U_2O_7$, that was accumulating as a byproduct of radium production. As the reduction temperature of uranium oxide by carbon is 1490° C., the electric furnace is the only available means for performing the operation in quantity. The furnace used in the experiments outlined in this paper was a small tilting furnace of the Siemens type, with a steel bottom contact, a rammed magnesite bottom, and a graphite electrode of 4-in. diameter. The power input was 30 to 40 kw. Considerable reactance was connected in the circuit, giving a furnace voltage of 30 to 60 volts.

Preliminary experiments were directed to the production of a 15-per cent. alloy, as several steel manufacturers preferred to try this grade first. Sodium uranate was charged, with petroleum coke as a reducing agent. This 15-per cent. ferro-uranium contained 3 per cent. carbon, and it seemed impossible to reduce the percentage of carbon.

At the start, uranium exhibited one of its fundamental characteristics, its strong affinity for carbon and its tendency to form carbides. Also, in this early work it was found that no fluxing material except fluorspar could be used with sodium uranate; if either lime or silica was added, the uranium remained in the slag with practically no reduction. Here was exhibited its second characteristic, its strong affinity for oxygen, the effect of which is shown in all later work. Because of its high atomic weight, uranium does not hold a large quantity of oxygen in combination, but holds very closely the small quantity which is in combination.

Control of Carbon

Four series of experiments were made with the object of producing low-carbon ferro-uranium containing 15 to 20 per cent. uranium:

(1) By variation of carbon in the charge.

(2) By removal of carbon from high-carbon alloy, with iron oxide.

(3) By removal of carbon from high-carbon alloy, with uranium oxide.

(4) By use of silicon as a reducing agent.

1. The effect of attempting to control carbon in ferro-uranium by the quantity of carbon charged is shown in Table 19. Except in experiments 3 and 4, the heats were made by melting steel turnings, and then charging a mixture of sodium uranate, coke, and fluorspar. In experiments 3 and 4, the steel turnings were mixed with the rest of the charge.

The sodium uranate used in the experiments recorded in Tables 19 and 20 contained 70 per cent. U_3O_8, or 59 per cent. uranium. The analysis of the steel turnings was: carbon, 0.31; silicon, 0.098; sulfur,

0.14; phosphorus, 0.107 per cent. The charcoal contained 64.73 per cent. and the coke 83.26 per cent. fixed carbon.

These experiments indicate clearly that a low-carbon ferro-uranium

TABLE 19.—*Control of Carbon in Ferro-uranium by Variation of Reduction Carbon*

	1	2	3	4	5	6	7	8	9	10
Charge:										
Steel turnings, lb.	20.00	20.00	20.00	20.00	20.00	20.00	20.00	10.00	10.00	10.00
Sodium uranate, lb.	22.00	22.00	22.00	22.00	22.00	22.00	15.00	10.00	10.00	10.00
Charcoal, lb.	44.50	4.50	6.50	6.50	4.50					
Coke, lb.		6.00			1.00	5.00	10.00	5.00	4.00	3.00
Fluorspar, lb.	4.00	2.00			3.00	4.00	2.00	3.00	2.00	2.00
Ferro-uranium										
Uranium, per cent.	nil	2.84	nil	2.93	nil	nil	7.72	18.3	20.5	19.5
Carbon, per cent.	0.43	3.22	1.94	2.41	1.52	0.84	4.11	3.87	3.49	4.07
Silicon, per cent.	1.39	3.01	1.42	0.08	1.13	0.19	1.52	3.03	2.53	2.19
Sulfur, per cent.	0.227	nil	0.041	0.05	nil	0.073	nil	nil	nil	nil
Phosphorus, per cent.	0.19	0.109	0.177	0.081	0.143	0.143	0.057	0.056	0.11	0.106
Iron, per cent.						99.0				
Slag:										
Uranium, per cent.	4.44	2.52	37.7		39.3					
Iron, per cent.	6.15	5.49	6.51		11.35					
Length of run, min.	222.0	240.0	185.0	95.0	145.0	170.0	125.0	160.0	78.0	80.0
Weight ferro-uranium, lb.	17.0	20.0	10.0	8.2	19.0	16.6	16.0	7.0	13.5	13.0
Weight slag, lb.	15.0	1.5	12.0	11.0	22.5					
Recovery of uranium, per cent.		4.2		4.3			14.1	21.7	46.9	42.7
Kw-hr. per lb. ferro-uranium tapped	5.3		8.3	7.5	3.2	5.6	4.8	12.1	3.4	3.4
Carbon charged per lb. of uranium, lb.	0.22	0.57	0.33	0.33	0.28	0.31	0.41	0.70	0.56	0.42

cannot be made by regulating the amount of carbon charged, as is the case with some other ferro-alloys. Because of its high reduction temperature, the reduction of uranium from [redacted] uranate requires carbon

about 400 per cent. in excess of the amount theoretically necessary. Theoretically, reduction takes place according to the reaction:

$$Na_2U_2O_7 + 6C = 2U + Na_2O + 6CO.$$

Until, for every pound of uranium, there is 0.4 lb. of carbon in the charge, very little reduction of uranium occurs. From 0.4 lb. carbon to 0.7 lb. carbon per pound of uranium, the percentage of uranium in the alloy varies little, although the recovery at this point is only about 45 per cent. Allowing for 15 per cent. mechanical loss, giving a theoretical recovery of 60 per cent. of the uranium, it appears that recovery of uranium, with over 0.7 lb. carbon per pound of uranium in the charge, would not increase, other conditions being the same. Hence it seems that increased recovery of uranium depends on a more efficient reduction furnace, and a higher temperature. It is possible, also, that better results may be obtained with another slag than fluorspar, or without any slag at all. Silica would suggest itself, as it might form sodium silicate.

At 0.4 lb. carbon per pound of uranium in the charge, the percentage of carbon in the alloy begins to vary a little, remaining between 3.5 and 4.0 per cent., up to 0.7 lb. carbon per pound of uranium. The critical point in this case is also 0.4 lb. carbon per pound of uranium, which was the point at which the percentage of uranium in the alloy became fixed within certain limits.

Curves show that until there is 3.5 per cent. carbon in the ferro-uranium, there will be little absorption of uranium; above 3.5 per cent. carbon, the uranium content varies little. The results on this point are not complete, but in general it appears that the carbon has to be over 2.5 per cent. before the alloy will contain any uranium, and at 3.5 per cent. to hold any large amount of uranium.

2. Table 20 shows the results of attempting to decarburize ferro-uranium with iron scale and silica, and with iron scale alone. The ferro-uranium was first melted before charging the iron scale and silica. The scale contained 71.55 per cent. iron.

The results of Table 20 show conclusively that ferro-uranium cannot be decarburized by the use of iron scale, because the uranium has so great a tendency to form iron uranate that it all passes into a slag and out of the ferro-uranium. Some of the carbon, however, is removed by this method.

3. Two experiments were made on the decarburization of ferro-uranium by the use of uranium oxide. A charge of 15 lb. of ferro-uranium, containing 14.25 per cent. uranium and 3.17 per cent. carbon, was melted with 7 lb. of uranium oxide containing 83.5 per cent. U_3O_8. The metal from the furnace contained no uranium and 2.95 per cent. carbon. In the second experiment, a carbon-lined furnace was used, to eliminate any oxidizing effect of the magnesite. A charge of 15 lb. of

ferro-uranium, containing 13.65 per cent. uranium and 3.17 per cent. carbon, was charged with 6 lb. of uranium oxide containing 83.5 per cent. U_3O_8. The carbon-lined furnace would naturally tend to increase the carbon in the alloy, but uranium should remain in the metal under these conditions, if it would remain there under an oxide slag in any furnace. After keeping the metal in the furnace for nearly 2 hr., the tapped metal contained no uranium, and 3.98 per cent. carbon.

TABLE 20.—*Decarburizing Ferro-uranium With Iron Scale*

	11	12	13	14	15
Charge:					
Ferro-uranium, lb	16.0	17.50	12.2	14.0	16.0
Iron scale, lb	8.5	7.00	6.2	2.5	3.0
Silica sand, lb	4.5	4.75	3.3		
Composition of ferro-uranium charged:					
Uranium, per cent	7.72	6.81	8.53	8.85	5.32
Carbon, per cent	4.11	2.9	.....	2.66	4.09
Silicon, per cent	1.52	1.62	.....	1.93	1.60
Sulfur, per cent	nil	nil	.....	nil	nil
Phosphorus, per cent	0.057	0.181	.....	0.067	0.061
Composition of alloy produced:					
Uranium, per cent	nil	nil	nil	nil	nil
Carbon, per cent	3.31	2.52	2.79	3.51	1.66
Silicon, per cent	1.74	1.75	nil	0.68	0.89
Sulfur, per cent	nil	nil	nil	nil	nil
Phosphorus, per cent	nil	nil	nil	0.079	0.069
Composition of slag:					
Uranium, per cent	13.37	10.00	10.67		
Length of run, min	115.00	180.00	90.00	85.00	80.00
Weight of alloy, lb	17.00	18.50	13.50	8.00	14.00
Recovery of uranium	nil	nil	nil	nil	nil
Kw.-hr. per lb. alloy tapped	3.9	4.4	3.2	5.2	2.8

These two experiments showed conclusively that ferro-uranium cannot be decarburized by the use of uranium oxide, because the uranium is entirely removed from the alloy.

4. Table 21 covers experiments on the production of ferro-uranium with silicon as a reducing agent, according to the reaction:

$$Na_2U_2O_7 + 3Si = 2U + Na_2SiO_3 + 2SiO_2$$

The raw materials were of the same composition as those used in the experiments of Table 20, and the silicon metal contained 90 per cent. silicon.

Ferro-uranium cannot be made by the use of silicon as a reducing agent, with either an efficient recovery of uranium or a low-silicon content of the alloy.

TABLE 21.—*Silicon Reduction of Ferro-uranium*

	16	17	18	19	20
Composition of charge:					
Steel turnings, lb.	10.0	10.0	10.0	10.0	10.0
Sodium uranate, lb.	10.0	10.0	10.0	10.0	10.0
Silicon, lb.	6.0	4.5	3.5	3.5	2.5
Fluorspar, lb.	2.0	2.0	2.0	2.0	2.0
Composition of metal:					
Uranium, per cent	2.95	4.01	1.6	12.3	9.47
Silicon, per cent	23.62	17.58	14.44	14.83	14.81
Carbon, per cent	0.25	0.31	1.22	1.64	0.70
Length of run, min.	135.00	120.00	100.00	120.00	105.00
Weight ferro-uranium, lb.	13.20	13.00	11.00	9.0	4.5
Recovery of uranium, per cent.	6.40	8.80	2.80	18.70	7.20
Kw.-hr. per lb. ferro-uranium tapped	4.60	5.00	4.80	8.30	13.30

Use of Uranium Oxide Instead of Sodium Uranate

Two experiments were made to determine the difference as to recovery of uranium and carbon content of ferro-uranium made from sodium uranate and from uranium oxide; the results are shown in Table 22. The sodium uranate contained 78 per cent. U_3O_8 or 65 per cent. uranium; the uranium oxide contained 82 per cent. U_3O_8, or 68.8 per cent. uranium. The coke analyzed 70 per cent. fixed carbon; the steel turnings, 0.31 per cent. carbon; and the steel stay-bolts, 0.11 per cent. carbon. Except for about two-thirds of the fluorspar, all of the charge was added as soon as the steel was melted.

TABLE 22.—*Uranium Recovery and Carbon Content with Sodium Uranate and with Uranium Oxide*

	21.—From Sodium Uranate	22.—From Uranium Oxide
Composition of charge:		
Steel turnings, lb.	10.00	10.00
Steel stay-bolts, lb.	.....	10.00
Sodium uranate, lb.	8.75	
Uranium oxide, lb.	.....	7.75
Coke, lb.	5.0	6.00
Fluorspar, lb.	7.0	8.0
Composition of metal:		
Uranium, per cent	25.10	29.20
Carbon, per cent	2.89	5.12(*a*)
Silicon, per cent	4.88	2.68
Length of run, min.	120.00	105.00
Weight ferro-uranium, lb.	14.20	14.25
Recovery of uranium, per cent.	63.80	76.20
Kw.-hr., per lb. ferro-uranium tapped	4.8	4.40

(*a*) Graphitic carbon, 1.53 per cent.

The uranium recovery, when using uranium oxide, is 12.4 per cent. higher than from sodium uranate. The ferro-uranium from heat No. 21 is low in carbon because of the high silicon, which throws carbon out of the alloy. No. 22 ferro-uranium is high in carbon, showing 1.53 per cent. graphitic carbon, because more carbon was charged than was needed for reduction. The carbon charged was figured in the same ratio to uranium as when sodium uranate was used, but uranium oxide here showed that it required less carbon per pound of uranium to give a good reduction.

From a commercial standpoint, all of these experiments were a failure. A low-carbon alloy could not be made by variation of the carbon in the charge, because there was such a low recovery of uranium, both when the amount of carbon charged was about the theoretical amount required, and when the carbon in the ferro-uranium was low. The carbon could not be removed from a high-carbon alloy with either iron oxide or uranium oxide, because the uranium in the alloy oxidized out of the metal and went into the slag, leaving pig iron in the furnace. Silicon reduction gave such a low recovery of uranium that its results were valueless. These negative results were caused by the two predominant characteristics of uranium which have been mentioned. Both analytical and microscopic examinations indicated that when there was enough carbon present to form the double carbide of iron and uranium, $Fe_3C.U_2C_3$, the ferro-uranium held the uranium, and only then. The net result of this work was the production of a ferro-uranium containing 15 to 30 per cent. uranium and 2.5 to 4 per cent. carbon.

Tests on Production of Uranium Metal

As it did not seem possible to produce a ferro-uranium containing 30 per cent. uranium with less than 3 per cent. carbon, experiments were made on production of as high-grade a uranium alloy as possible, with a view to decreasing the ratio of carbon to uranium, while holding the carbon to less than 5 per cent., and uranium over 85 per cent.

Experiments were also made on the reduction of uranium oxide with sodium, aluminum, and silicon as reducing agents.

Six ounces of metallic sodium was placed with 1 lb. of uranium oxide, containing 91 per cent. U_3O_8, in a graphite crucible heated by gas. After 18 hr., the crucible was removed; there was no reduction of uranium and the crucible had been destroyed. The experiment was repeated for 5 hr. with negative results.

An attempt was made to reduce 1 lb. of uranium oxide with 2 oz. of coke in a gas-fired crucible. Heating was continued for 18 hr. The crucible was destroyed and there was no reduction.

A mixture of aluminum, silicon, and carbon was tried as a reducing

agent, using Siemens furnace with carbon bottom and magnesite sides. A charge of 15 lb. of fluorspar was melted, and then a mixture of 20 lb. uranium oxide, 10 lb. petroleum coke, 1 lb. aluminum shot, and 0.75 lb. silicon metal was added, followed, toward the end, by 15 lb. of fluorspar. The length of the run was 3 hr. The furnace was allowed to cool and a small button of metal was removed, which contained: uranium, 11.85; vanadium, 8.66; iron, 52.16; carbon, 1.12; silicon, 13.25, and aluminum, 12.96 per cent. During the experiment there appeared to be a reduction, but it evidently was aluminum burning.

TABLE 23.—*Production of Uranium Metal*

	23	24	25	26
Charge:				
Uranium oxide, lb.	15.00	20.00	20.00	20.00
Petroleum coke, lb.	5.75	3.00	2.00	2.00
Metal:				
Uranium, per cent	85.50	88.10	93.00	88.00
Carbon, per cent	9.31	7.49	4.39	3.67
Silicon, per cent	0.29	1.48	1.43	2.47
Iron, per cent	0.82	2.43	1.35	1.03
Vanadium, per cent	3.64	1.82	1.21	1.14
Phosphorus, per cent	.....	.....	0.051	
Sulfur, per cent	.....	.....	0.13	
Residue:				
Uranium, per cent	57.76	19.76		
Carbon, per cent	12.32	2.65		
Silica, per cent	.....	6.62		
Iron, per cent	.....	3.42		
CaO, per cent	.....	12.67		
MgO, per cent	.....	6.57		
Length of run, min.	105.00	280.00	133.00	170.00
Weight metal, lb.	6.94	13.20	9.75	9.31
Recovery of uranium, per cent	49.30	76.40	56.30	50.80
Uranium content, (by diff.)per cent.	85.94	86.78	91.44	91.69
Kw.-hr. per lb. metal tapped	12.90	11.10	7.60	9.60
Carbon charged per lb. uranium, lb.	0.42	0.16	0.10	0.10

In another experiment in the same furnace, 15 lb. fluorspar was melted, and the following charge added slowly: 20 lb. uranium oxide, 2.8 lb. aluminum shot. The length of the run was 2½ hr. On cooling the furnace, 1 lb. of metal was obtained which contained no uranium. There was a violent reaction when charging, probably due to burning of aluminum.

Subsequent to these preliminary experiments, ferro-uranium was made by William Y. Bleakley, using aluminum as a reducing agent.

He states that the uranium content is easily controlled, and that the carbon can be held under 1 per cent. and the aluminum from 3 to 5 per cent.

Table 23 contains the results of experiments on the production of uranium metal, using carbon as a reducing agent, in Siemens furnace with magnesite walls and carbon bottom. The metal was allowed to cool in the furnace. In experiments No. 23 and 24, the uranium oxide contained 91 per cent. U_3O_8, but in the other runs it contained 94.7 per cent. U_3O_8. The analysis of the petroleum coke was: fixed carbon, 85; ash, 3.81; and volatile matter, 11.89 per cent.

The results in Table 23 indicate that a uranium metal can be made containing over 90 per cent. uranium and from 3.5 to 4 per cent. carbon. The production of this low-carbon metal, with the minimum amount of carbon charged, depends entirely on manipulation of the electric furnace, a very high temperature being necessary.

The product in each case, except No. 23, is uranium metal or an alloy of uranium carbide and uranium metal. In No. 23, the product, as shown under the microscope, is almost entirely uranium carbide. No. 24 is an alloy of uranium carbide and uranium, with a considerable proportion of carbide. Products No. 25 and 26 are almost all uranium, with a very small amount of uranium carbide. The silicon and iron can be kept at about 1.5 per cent. each, and the vanadium at 1.25 per cent., with the oxide used.

The uranium recovery in a single operation will be from 50 to 60 per cent. The material left in the furnace is a mixture of pure oxide and carbon, which can be charged again as soon as its carbon content is determined. In this way, the ultimate recovery of uranium is high, well over 85 per cent. The power consumption is high, because pure uranium oxide is being reduced. It averages about 8 kw.-hr. per pound of metal.

Further experiments were made, Table 24, in an attempt to produce a lower-carbon uranium metal. The furnace was lined both on sides and bottom with magnesite, the bottom contact being a ½-in. carbon rod. In each run the charge was added slowly, at the rate of ¾ lb. of charge every 5 min.; the metal was left in the furnace until cool. The petroleum coke contained 85 per cent. fixed carbon.

The experiments of Table 24 do not permit different conclusions from those previously made regarding the production of uranium metal. The last series of experiments indicates that when an attempt is made to produce an alloy containing under 1 per cent. carbon, the greater part of the product is UO_2 mixed with uranium carbide and a little metallic uranium. The products obtained in experiments Nos. 27 to 31, inclusive, were of this character, judging from the uranium content and the non-metallic appearance of the material.

When the carbon was from 1 to 2 per cent., the product was a mixture of uranium metal, uranium carbide, and uranium oxide. Of the total uranium, the amount present as oxide varies from 12 to 40 per cent.

TABLE 24.—*Production of Uranium Metal*

	27	28	29	30	31	32	33	34	35	36	37
Composition of charge:											
Uranium oxide (80.5% U) lb.	20.0	20.0	10.4								
Uranium oxide (69.5% U) lb.			20.0								
Uranium oxide (72.5% U) lb.				35.0		18.75					
Uranium oxide (73.0% U) lb.							20.0	20.0	20.0	45.0	47.0
Uranium oxide (87.7% U) lb.					24.5						
Uranium slag (57.7% U) lb.							20.75				
Uranium oxide (84.8% U) lb.				10.3							
Petroleum coke	1.87	1.81	2.81	4.5	3.0	2.38	5.5	3.5	3.5	7.87	7.87
Composition of product:											
Uranium, per cent.	84.8	83.7	81.5	85.5	87.5	88.8	85.28	86.73	79.0	83.4	85.8
Carbon, per cent.	0.63	0.45	tr.	0.49	0.9	1.47	2.06	3.95	4.70	1.4	4.52
Silicon, per cent.	0.73	1.08	0.29	0.09	1.27	0.23	0.96	0.42	0.94	0.36	0.36
Iron, per cent	0.47	0.16	tr.	tr.	1.15	tr.	2.83	1.25	4.19	0.73	1.03
Vanadium, per cent.	0.97	1.15	0.33	1.95	1.85	2.16	2.96	3.91	3.22	3.03	2.82
Length of run, min.	150.0	165.0	155.0	360.0	180.0	120.0	300.0	195.0	195.0		
Weight metal, lb.	29.5	7.75	17.0	17.25	8.5	9.83	19.0	11.3	18.0	17.25	13.87
Recovery of uranium, per cent.	47.4	38.1	57.9	43.5	36.2	63.7	50.5	67.1	97.3		
Oxygen consumed on ignition						11.49	13.42	17.56	18.03	10.62	17.52
Uranium:											
present as metal, per cent.						37.6	58.5	39.8	21.4	49.8	33.9
present as carbide, per cent.						21.8	29.5	60.2	78.6	22.1	66.1
present as oxide, per cent.						40.6	12.0	nil	nil	28.1	nil
Kw.-hr. per lb. of product.	9.0	10.6	6.3	10.0	12.7	6.2	7.6	18.8	11.8		
Carbon charged per lb. uranium, lb.	0.1	0.09	0.104	0.15	0.19	0.15	0.17	0.2	0.2	0.2	0.19

The product was high in uranium but low in impurities, also showing that reduction was not complete.

When the carbon in the product is from 3.5 to 5 per cent., all of the uranium is present as carbide or metal. Of the total uranium present, 25 to 50 per cent. is metal and the rest carbide. These conclusions were

confirmed by microscopic examination, in addition to determination of oxide by analysis.

An interesting feature developed in these experiments was the ease with which metallic uranium oxidizes. In working with a charge which produced uranium, or metal and carbide, if the button of sponge metal was removed from the furnace while still red hot, and exposed to the air, it immediately oxidized to black uranium oxide, a 25-lb. button being converted completely to oxide in less than 5 minutes.

Manufacture of Uranium Metal

The materials used in the manufacture of uranium metal are uranium oxide and petroleum coke. Both of these materials should be low in impurities, especially in iron and silica. The most suitable furnace for the production of uranium metal is a stationary Siemens electric furnace, having the crucible bottom as one electrode and a single suspended electrode. The lining is ground, dead-burned magnesite, rammed into place with pitch as a binder, and the bottom electrical connection is made by a small graphite rod buried in the magnesite. Fireclay tiles of hexagon cross-section also may be used for a furnace.

Uranium oxide and petroleum coke are mixed in proper proportions and charged slowly into the furnace. To secure the proper temperature, the furnace must arc all the time. Regular charging is necessary. The uranium metal is allowed to form a sponge on the bottom of the furnace, which gradually builds up. Directly beneath the electrode, this sponge is pasty, but not fluid. The charge forms its own furnace lining. When a run is finished, the button of uranium metal is removed from the furnace. The unreduced residue is re-treated with another charge.

The reaction involved is a simple carbon reduction, and the product consists of a mixture of metallic uranium and uranium carbide.

$$UO_2 + 2C = U + 2CO.$$
$$U_3O_8 + 8C = 3U + 8CO.$$
$$3UO_2 + 5C = U + U_2C_3 + 2CO.$$
$$U_3O_8 + 11C = U + U_2C_3 + 8CO.$$

A typical charge for the production of uranium metal would be: uranium oxide, 20 lb.; petroleum coke, 2 lb., the oxide containing 94.7 per cent. U_3O_8, and the petroleum coke, 85 per cent. fixed carbon.

Making a recovery of 56 per cent. of the uranium, 100 lb. of this charge will produce 44 lb. of uranium metal. By resmelting the residue, practically all of the uranium can be recovered except the small amount lost mechanically.

The uranium metal made from this charge contained: uranium, 93.0; carbon, 4.39; silicon, 1.43; iron, 1.35; vanadium, 1.31; phosphorus, 0.051; sulfur, 0.13 per cent.

Addition of Uranium Metal to Steel

Steel manufacturers who had been trying ferro-uranium for making high-speed steel reported that they could not make uranium stay in the steel, getting a recovery from zero to 50 per cent., when using the 15 to 30-per cent. grade of ferro-uranium. Uranium metal was submitted with the expectation that a 75 per cent. recovery would be made, on addition to steel, but they reported the same results as before.

A series of tests was made on the production of a uranium steel by the addition of uranium metal, Table 26. The steel was made, in each case, in a tilting Siemens furnace lined with magnesite. Stay-bolts were first melted with a slag of lime, iron scale, and fluorspar; the slag was skimmed, and a second charge of lime and fluorspar was added. The first slag was

TABLE 25.—*Analyses of Materials Used in Making Uranium Steel from Uranium Metal*

Stay-bolts	Per Cent.
C	0.11
Mn	nil
Si	0.14
P	0.148
S	0.014

Lime	Per Cent.
CaO	61.45
MgO	11.34
SiO_2	2.48
Vol.	24.42

Iron Scale	Per Cent.
Fe	71.55

Petroleum Coke	
Fixed C	84.8
Volatile	11.39
Ash	3.81

Fluorspar No. 1	Per Cent.
CaF_2	89.43
SiO_2	0.21
Al_2O_3	4.28
Fe_2O_3	nil
$CaCO_3$	1.21

Fluorspar No. 2	Per Cent.
CaF_2	63.04
SiO_2	5.80

Uranium Metal	Per Cent.
U	80.28
C	4.43
Si	0.60
Fe	2.39
V	3.24

Ferromanganese.. 80 per cent. Mn
Ferrosilicon..... 50 per cent. Si
Silica, high-grade sand
Aluminum, shot

for keeping down the carbon and phosphorus, and the second to take care of the sulfur. When the second slag was melted, petroleum coke was added. The slag then turned white; if it became black again, more coke was added. Before pouring, ferromanganese and ferrosilicon were charged into the furnace, and shortly afterward the uranium was added, if the test called for its addition in the furnace. In some runs, only the second type of slag was used, and in some runs this slag contained silica as well as lime and fluorspar. Aluminum was sometimes added, in the ladle. To test the effect of adding uranium metal with a slag of fluorspar instead of lime and fluorspar, in some experiments the second slag was skimmed and was followed by a slag of fluorspar. If necessary, coke

TABLE 26.—*Uranium Steel by Addition of Uranium Metal*

	38	39	42	43	44	45	49	50	51	52	53	55	56	57	58
Charge:															
Stay-bolts, lb.	20.0	20.00	20.0	20.0	20.0	20.0	20.0	20.0	20.0	20.0	20.0	20.0	20.0	20.0	20.0
1st Slag:															
Lime, lb.	.....	.....	1.5	1.5	1.5	1.5	1.5	1.5	1.5	1.5	1.5	1.5	1.5	1.5	1.5
Iron scale, lb.	.....	.....	0.5	1.0	1.0	1.0	1.0	1.0	1.0	1.0	1.0	1.0	1.0	1.0	1.0
Fluorspar No. 1, lb.	.....	.....	0.25												
Fluorspar No. 2, lb.	.....	.....	.....	0.375	0.375	0.375	0.375	0.375	0.375	0.375	0.375	0.375	0.375	0.375	0.375
2d Slag:															
Lime, lb.	2.0	2.0	2.0	2.0	2.0	2.0	2.0	2.0	2.0	2.0	2.0	2.0	2.0	2.0	2.0
Silica, lb.	0.375	0.375	0.375												
Fluorspar No. 1, lb.	1.0	1.0	1.0												
Fluorspar No. 2, lb.	.....	.....	.....	2.0	2.0	2.0	2.0	2.0	2.0	2.0	2.0	2.0	2.0	1.5	1.5
3d Slag:															
Fluorspar No. 2, lb.	.....	.....	.....	4.0	4.0	3.0									
Petroleum coke, grams.	84.0	56.0	84.0	84.0	112.0	112.0	84.0	84.0	84.0	112.0	140.0	112.0	112.0	162.0	162.0
Ferromanganese, grams.	168.0	168.0	70.0	25.0	25.0	25.0	30.0	30.0	30.0	30.0	30.0	30.0	30.0	40.0	40.0
Ferrosilicon, grams.	.....	.....	56.0	5.0	5.0	5.0	5.0	5.0	5.0	5.0	5.0	5.0	5.0	10.0	10.0
Aluminum, grams.	.....	.....	1.5	1.0	1.0	1.0									
Uranium metal, grams.	28.0	28.0	28.0	142.0	142.0	115.0	170.0	170.0	170.0	170.0	170.0	170.0	170.0	170.0	170.0
Analysis steel:															
Carbon, per cent.	1 50	1.11	1.86	0.31	0.90	0.47	0.24	0.24	0.12	0.75	0.44	0.29	0.32	1.31	0.50
Manganese, per cent.	0.75	1.84	0.08	0.10	0.03	0.10	0.24	0.17	0.12	0.01	0.25	0.13	0.25	0.46	0.39
Silicon, per cent.	1.18	0.58	0.98	0.06	0.197	0.193	0.047	0.05	0.03	0.28	0.24	0.19	0.14	0.74	0.54
Phosphorus, per cent.	0.134	0.155	0.068	0.092	0.061	0.069	0.07	0.05	0.025	0.19	0.13	0.13	0.11	0.13	0.08
Sulfur, per cent.	0.03	0.016	0.044	0.03	0.038	0.027	0.02	0.03	0.04	0.008	0.016	0.016	0.014	0.005	0.008
Uranium, per cent.	nil	nil	nil	nil	nil	nil	0.18	nil	nil	0.14	0.07	0.04	0.07	0.24	0.27
Uranium added in.	F	F	F	F	F	F	L	F	L	F	F	F	F	L	L
Slag on steel when adding uranium.	2d	2d	2d	3d	3d	3d	none	none	none	2d	2d	2d	2d	none	none
Uranium in furnace, min.	5.0	8.0	6.0	1.0	21.0	15.0	0.0	2.0	0.0	½	1.0	3.0	0.0	0.0	0.0
Recovery of uranium, per cent.	0.0	0.0	0.0	0.0	0.0	0.0	11.6	0.0	0.0	9.0	4.5	2.5	4.5	15.4	17.4

F—Furnace. L—Ladle. In experiments 40, 41, 46, 47, 48, omitted from table for lack of space, the recovery of uranium was 0. In experiment No. 54, omitted, the conditions and results were practically identical with those of No. 53, except that the uranium was in the furnace for 2 min.

was thrown on this slag. A graphite crucible was used as a ladle. The steel was poured into molds of 2- or 3-in. square cross-section, or into pigs.

The composition of the raw materials used is shown in Table 25. The uranium metal and all of the ferro-alloys were crushed to about pea size.

In experiments No. 38 to 42, inclusive, the uranium metal was added to the steel, in the furnace, after addition of ferromanganese and ferrosilicon, and was left there for 5 to 10 min.; no uranium was found in the steel. The lime-fluorspar-silica slag was on the steel while the uranium was in the furnace.

In experiments No. 43 to 49, inclusive, the second slag was skimmed and a slag of fluorspar was placed on the steel. Then the ferromanganese and ferrosilicon were charged, followed by the uranium metal. The steel was then left in the furnace for from 1 to 15 min. before pouring. Even with the fluorspar slag, no uranium was found in the steel; but most of it was in the slag, as shown in the following analyses. The character of the steel to which the uranium was added is shown also, the carbon sample having been taken before addition of uranium metal.

No.	Uranium in Slag, Per Cent.	Carbon in Steel, Per Cent.
43	2.01	0.13
44	1.34	0.16
45	2.01	0.33
46	2.35	0.08
47	3.81	0.09
48	3.65	0.16

The uranium content of the slag indicates that the uranium metal was melting partly, but slagged at once if any slag was present.

As a result of the slag analyses, a test was made in which the uranium was thrown into the molten stream while the steel was being poured from the furnace into the ladle; before pouring, the slag was skimmed from the steel. This run, No. 49, was the first to yield steel containing uranium. In experiments No. 49 to 59, the uranium metal was added in the ladle, in the furnace after skimming the slag, or in the furnace with the slag, a variable length of time elapsing after the addition of uranium. The best results were obtained in No. 57 and 58, in which the slag was skimmed, the steel poured, and the uranium thrown into the stream of metal as it entered the ladle. In No. 51, the uranium metal was placed in the ladle before pouring, with negative results, the uranium forming a cake which did not melt.

Uranium metal is not a satisfactory agent for the addition of uranium to steel, for two reasons:

(1) It has such a high melting point that if it is added to the steel in the furnace just before pouring, or in the ladle, all of the metal is not melted, and only a comparatively small proportion enters the steel.

(2) When it is left in the steel bath for a period long enough to melt it, it passes into the slag and oxidizes so easily that no uranium is recovered in the steel; it all slags off.

The first conclusion is based on the results obtained in tests No. 49 to 58, inclusive, in each of which a considerable amount of uranium metal was found untouched after pouring, whether it was added in the furnace or in the ladle.

TABLE 27.—*Raw Materials for Making Ferro-uranium*

Steel Turnings	Per Cent.
Carbon	0.20
Manganese	0.74
Silicon	0.118
Phosphorus	0.092
Sulfur	0.101

Petroleum Coke No. 1	Per Cent.
Carbon	85.44
Volatile	11.63
Ash	2.93

Fluorspar	Per Cent.
CaF_2	89.43
SiO_2	0.21
$CaCO_3$	1.21
Al_2O_3	
Fe_2O_3	4.28

Uranium Oxide No. 1	Per Cent.
Uranium	73.7
SiO_2	2.2

Uranium Oxide No. 2	Per Cent.
U	73.33
V	5.07
SiO_2	2.39

	Sodium Uranate 1 Per Cent.	2 Per Cent.	3 Per Cent.	4 Per Cent.
U_3O_8	65.40	81.40	67.70	60.70
V_2O_5	5.47	1.20	3.04	2.71
FeO	1.06	Tr.	Tr.	
SiO_2	1.06	2.17	3.83	8.87
Al_2O_3	4.77	0.70	5.50	

Uranium Slag No. 1	Per Cent.
Uranium	57.0
Fe_2O_3	23.6
SiO_2	3.5

Uranium Slag No. 2	Per Cent.
Uranium	44.1
Carbon	5.88
Vanadium	1.04
Iron	10.76
SiO_2	3.14

Uranium Slag No. 3	Per Cent.
Uranium	41.84
Carbon	3.50
Iron	12.96
SiO_2	2.10

The second conclusion is based on all of the experiments up to No. 49, no uranium having been found in the steel when the metal was left in the furnace for any length of time. The character of the slag on the steel at the time the uranium is added apparently has no effect on the recovery of uranium; hence the slag can be selected to produce the best steel.

It is not probable that a recovery of over 25 per cent. of the uranium can be expected by the use of uranium metal, and this will be difficult to attain when it is desired to add uranium in large amounts. To date, the use of uranium metal does not offer much encouragement for a high

recovery of uranium, and only when it is added to the steel in the ladle is there any great degree of success.

Tests on Production of 50-per cent. Ferro-uranium

Because uranium could not be added to steel as uranium metal, experiments were made on the production of a 50-per cent. ferro-uranium. This alloy would have a melting point between 1500 and 1700° C., and its iron should tend to hold the uranium away from the slag when melted.

Three series of experiments were made: Table 28 using uranium oxide, Table 29 using sodium uranate, and Table 30 using the unreduced residues and slags from the runs of Table 24. In Table 27 are given analyses of raw materials charged.

TABLE 28.—*Production of Ferro-uranium from Uranium Oxide and Slag*

	59	60	61	62	63	64
Charge:						
Steel turnings, lb	10.0	20.0	15.0	15.0	22.5	22.5
Fluorspar, lb	13.0	37.0	25.0	15.0	15.0	20.0
Petroleum coke, lb	14.3	25.5	13.0	7.0	7.87	10.5
Uranium slag No. 1, lb	52.0	108.0	28.0			
Uranium oxide No. 2, lb	.....	.....	26.0			
Uranium oxide No. 3, lb	.....	.....	.....	40.0	60.0	60.0
Analysis, ferro-uranium:						
Uranium, per cent	56.50	52.87	60.40	50.40	36.40	37.90
Carbon, per cent	4.47	4.74	4.85	4.95	4.24	4.91
Silicon, per cent	2.54	2.33	1.19	1.16	1.79	2.00
Vanadium, per cent	2.31	2.15	1.69	3.25	4.81	4.29
Iron, per cent	33.80	37.20	29.36	39.33	46.20	50.10
Graphite, per cent	0.40	0.07	nil	nil	nil	nil
Phosphorus, per cent	0.072	0.053	0.099	0.066	0.085	
Sulfur, per cent	nil	nil	nil	nil	nil	
Length run, min	405.0	365.0	215.0	170.0	240.0	250.0
Weight ferro-uranium, lb	21.0	16.5	40.0	32.0	12.0	11.5
Power consumption per lb., kw.-hr.	8.5	3.4	3.3	2.9	.....	11.6
Uranium recovery, per cent	48.9	55.5	63.1	55.2		

Additional metal cleaned from the furnace was as follows:

Heat No	59	60	61	63
Weight metal, lb	4.87	24.00	20.00	2.87
Uranium, per cent	54.50	56.81	59.38	45.15
Carbon, per cent	6.44	6.88	5.77	4.39
Silicon, per cent	2.64	2.06	2.04	
Vanadium, per cent	2.21	2.10	1.82	
Iron, per cent	33.20	32.70	30.20	
Graphite, per cent	1.75	1.26	1.38	
Phosphorus, per cent	0.069	0.056	0.052	
Sulfur, per cent	nil	nil	nil	

The general procedure in all of these experiments, except No. 59 and 60, was to mix the iron turnings with the rest of the charge and add it

TABLE 29.—*Production of Ferro-uranium from Sodium Uranate*

	68	69	70	71	72	73	74	75	76	78	79	80	81	82	83
Charge:															
Steel turnings, lb.	15.0	15.0	15.0	15.0	15.0	15.0	15.0	15.0	15.0	15.0	15.0	15.0	15.0		15.0
Petroleum coke, lb.	14.0	14.0	10.5	9.3	14.0	14.0	11.0	11.0	14.0	15.0	15.0	15.0	15.0		15.0
Fluorspar, lb.	15.0	.....	.....	10.0	15.0	30.0	40.0	30.0	33.0	36.0	36.0	36.0	36.0		36.0
Sodium uranate No. 1, lb.	50.0	40.0	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....		.....
Sodium uranate No. 2, lb.	.....	10.0	50.0	34.0	.....	.....	.....	.....	.....	.....	.....	.....	.....		.....
Sodium uranate No. 3, lb.	.....	.....	.....	.....	50.0	50.0	40.0	40.0	50.0	.....	.....	.....	.....		.....
Sodium uranate No. 4, lb.	.....	.....	.....	.....	.....	.....	.....	.....	.....	60.0	60.0	60.0	60.0		60.0
Analysis metal:															
Uranium, per cent	41.6	48.3	51.9	46.7	45.1	49.5	32.2	32.28	5.88	49.9	39.0	41.2	.....	.....	.....
Carbon, per cent	5.22	5.56	5.03	6.06	5.5	5.05	4.63	4.63	2.84	5.5	5.4	4.95	.....	.....	.....
Silicon, per cent	0.60	1.16	0.8	1.18	2.14	1.56	1.62	1.62	4.54	3.97	2.6	3.34	.....	.....	.....
Vanadium, per cent	3.35	2.55	0.86	0.81	1.52	1.74	1.88	1.88	1.21	11.9	6.72	18.1	.....	.....	.....
Aluminum, per cent	0.55	1.08	0.55	0.12	0.06	0.13	.....	.....	.....	.....	.....	.....	.....	.....	.....
Length run, hr.	4.0	4.0	3.2	2.0	2.3	5.6	3.5	3.5	2.6	4.5	4.5	8.0	4.0		8.0
Weight ferro-uranium, lb.	36.0	20.0	9.0	16.0	12.0	45.0	none	19.0	8.0	23.0	.....	26.0	none		none
Kw.-hr. per lb. ferro-uranium	3.7	6.9	12.2	4.7	13.5	4.4	.....	8.3	19.0	12.3	.....	.....	.....	.....	.....
Recovery of uranium, per cent	47.3	33.7	13.5	23.2	18.9	77.6	.....	26.5	1.7	37.1	.....	.....	.....	.....	.....

slowly to the furnace at the rate of 2 lb. of charge every 5 min. After the last addition of mixed charge, some fluorspar was charged and the fur-

nace operated 15 min. before pouring. In No. 59 and 60, the steel was melted before addition of the rest of the charge.

From the results of Table 28, it is evident that a ferro-uranium containing 40 to 60 per cent. uranium and under 5 per cent. carbon can be made about as readily as a lower grade alloy. The silicon can be kept generally under 2, but at times is as high as 3 per cent. When the 50-per cent. alloy contains about 5 per cent. carbon, it contains little or no graphite.

The power consumption varies from 3 to 5 kw.-hr. per pound of alloy. A recovery of 50 to 60 per cent. of the uranium charged can be

TABLE 30.—*Production of Ferro-uranium from Slag*

	84	85	86	87	88	89	90*	91	92
Charge:									
Steel turnings, lb	14.0	15.0	15.5	13.5	17.0	12.0	.....	13.0	11.5
Fluorspar, lb	.....	15.0	15.0	15.0	20.0	15.0	.....	15.0	15.0
Petroleum coke, lb	7.0	7.5	7.5	7.5	10.5	7.5	.....	7.5	7.5
Uranium oxide, No. 2, lb	40.0	.....	.....	.....	3.5	.....	.....	.....	.....
Uranium slag, No. 2, lb	.....	60.0	60.0	60.0	34.5	.....	.....	.....	.....
Uranium slag, No. 3, lb	.....	.....	.....	.....	44.0	63.0	.....	63.0	63.0
Analysis metal:									
Uranium, per cent	33.3	43.6	38.29	40.1	39.35	40.0	44.9	42.8	41.9
Carbon, per cent	4.68	4.87	3.8	4.9	4.24	4.52	5.58	4.97	5.02
Silicon, per cent	2.03	3.10	5.36	3.92	6.42	4.09	2.01	2.40	2.02
Vanadium, per cent	4.49	2.72	2.47	2.34	1.82	2.60	2.6	2.08	2.21
Iron, per cent	55.44	42.0	40.46	42.17	37.1	38.6	37.6	43.4	45.85
Length run, min	175	275	280*	225	330	220	.....	260	235
Weight ferro-uranium, lb	5.12	43.75	35.5	37.5	29.0	24.0	29.0	40.0	27.5
Kw.-hr. per lb	.....	3.6	4.8	3.7	4.7*	3.4*	.....	3.7	5.0
Uranium recovery, per cent	.....	72.2	55.3,	57.0	48.0*	61.0*	.....	65.3	44.0

* No. 90 was metal removed from the furnace bottom after runs No. 88 and 89 and is distributed evenly between them in figuring power consumption and recovery.

made in a single operation. The low recovery in some of the experiments was due to attempts to keep the carbon lower than usual by charging less coke.

The experiments of Table 29 show an average recovery of 31 per cent. of the uranium charged, and a power consumption of 10.5 kw.-hr. per pound of alloy tapped. The results are very irregular, and the carbon has a tendency to be higher than when the alloy is made by reduction of oxide. With the exception of vanadium, the ferro-uranium produced is no higher in impurities than the alloy made from uranium oxide. In general, the use of sodium uranate rather than oxide results in a much lower average recovery of uranium, and a much higher power consumption.

The production of ferro-uranium from slags and unreduced residues,

Table 30, shows about the same recovery and power consumption as when new uranium oxide is charged. The ferro-uranium produced is of a high grade, and cannot be distinguished from the products of Tables 28 and 29 in appearance or analysis.

Percentage Recovery of Uranium

It is now possible to get an idea of the total uranium recovery possible in making ferro-uranium.

	Actual Recovery in Experiments, Per Cent.	Estimated Recovery in Practice, Per Cent.
Ferro-uranium from sodium uranate	63.8	75
Ferro-uranium from uranium oxide	76.2	85
Ferro-uranium from sodium uranate with slags smelted once	87.6	91
Ferro-uranium from uranium oxide with slags smelted once	91.3	96

From these results, it appears that by smelting the slag, the ultimate recovery of uranium from uranium oxide is not much better, viewed from a commercial standpoint, than from sodium uranate. The use of the oxide, however, may result in so high a recovery in one operation that resmelting the slags will not be necessary. Also, the use of sodium uranate is certain to result in silicon in the alloy, especially in the ferro-uranium made by smelting slags. Uranium oxide must be used to give an alloy low in silicon, in which the only important impurity is carbon, from 3 to 4 per cent.

Table 31.—*Analyses of Materials Used in Making Uranium Steel with Ferro-uranium*

Stay-bolts	Per Cent.
Carbon	0.10
Manganese	0.05
Silicon	0.52
Phosphorus	0.028
Sulfur	0.021

Very variable.

Steel Turnings	Per Cent.
Carbon	0.3
Manganese	0.74
Silicon	0.118
Phosphorus	0.092
Sulfur	0.101

Ferro-uranium No. 1	Per Cent.
Uranium	25.1
Carbon	3.78
Silicon	5.73

Nickel-uranium No. 2	Per Cent.
Uranium	20.83
Nickel	62.6
Carbon	0.36
Silicon	9.6
Iron	1.5
Aluminum	2.07

Ferro-uranium No. 3	Per Cent.
Uranium	25.1
Carbon	2.89
Silicon	4.88
Phosphorus	0.074
Sulfur	nil

Ferro-uranium No. 4	Per Cent.
Uranium	24.85
Carbon	4.14
Silicon	4.83

Ferro-uranium No. 5	Per Cent.
Uranium	29.2
Carbon	5.12
Silicon	2.68

Ferro-uranium No. 6	Per Cent.
Uranium	19.5
Carbon	3.9
Silicon	2.59

Ferro-uranium No. 7	Per Cent.
Uranium	37.55
Carbon	3.54
Silicon	2.32
Vanadium	1.18

TABLE 32.—*Uranium Steel by Addition of Ferro-uranium*

	95	96	97	98	99	100
Charge:						
Stay-bolts, lb.	20.0	20.0	20.0	20.0	20.0	20.0
Steel turnings, lb.						
1st slag:						
Lime, lb.	1.5	1.5	1.5	1.5	1.5	1.5
Iron scale, lb.	1.0	1.0	1.0	1.0	1.0	1.0
Fluorspar, lb.	0.375	0.375	0.375	0.375	0.375	0.375
2d slag:						
Lime, lb.	2.0	2.0	2.0	2.0	2.0	2.0
Fluorspar, lb.	1.5	1.5	1.5	1.5	1.5	1.5
Petroleum coke, grams.	162.0	112.0	112.0	112.0	112.0	112.0
Ferromanganese, grams.	40.0	40.0	40.0	40.0	40.0	40.0
Ferrosilicon, grams.	.....	10.0	10.0	5.0	5.0	
Ferro-uranium, No. 1, lb.	3.06	4.0	2.38	1.18	1.18	1.62
Nickel-uranium, No. 2, lb.	0.69					
Ferro-uranium, No. 3, lb.	.....	.....	.....	.....	.....	.....
Ferro-uranium, No. 4, lb.	.....	.....	.....	.....	.....	.....
Ferro-uranium, No. 5, lb.	.....	.....	.....	.....	.....	.....
Ferro-uranium, No. 6, lb.	.....	.....	.....	.....	.....	.....
Ferro-uranium, No. 7, lb.	.....	.....	.....	.....	.....	.....
Analysis steel:						
Carbon, per cent.	0.77	0.61	0.63	0.49	0.50	0.41
Manganese, per cent.	0.46	0.39	0.42	0.39	0.12	0.28
Silicon, per cent.	1.3	0.68	0.83	0.49	0.49	0.38
Phosphorus, per cent.	0.08	0.06	0.10	0.066	0.058	0.045
Sulfur, per cent.	0.011	0.011	0.011	0.009	0.003	0.006
Uranium, per cent.	2.55	2.18	1.81	1.42	1.69	1.02
Nickel, per cent.	2.07	.....	.....	.....	.....	.....
Vanadium, per cent.	.....	.....	.....	.....	.....	.....
Slag on steel when adding ferro-uranium.	none	none	none	none	none	none
Uranium in furnace, min.	0.0	0.0	0.0	0.0	0.0	0.0
Recovery of uranium, per cent.	62.8	48.8	65.3	100.0	118.0	53.1
Desired per cent. uranium in steel.	4.06	4.46	2.77	1.42	1.42	1.92

Addition of Ferro-uranium to Steel in Electric Furnace

Experiments were made on the production of uranium steel by the addition of ferro-uranium to molten steel. The steel was made in a tilting Siemens furnace lined with magnesite, by the same procedure described under the use of uranium metal. The ferro-uranium was added either in the ladle or in the furnace. If in the furnace, the addition followed that of ferromanganese and ferrosilicon. Sometimes the slag was skimmed before adding the ferro-uranium; at other times it was left in. The length of time the ferro-uranium remained in the furnace before pouring the steel was also varied, from immediate pouring to one minute. A third slag of fluorspar was not used in these experiments, the tests on uranium metal having shown that it did not increase the recovery.

Analyses of the steel turnings, stay-bolts and ferro-uranium used in the tests are shown in Table 31; other materials were the same as in the

TABLE 32.—*Uranium Steel by Addition of Ferro-uranium—(Continued)*

102	104	105	106	107*	110	113	115	116	117	118	120
20.0	20.0	20.0	20.0	20.0	20.0	20.0	20.0	20.0	20.0	20.0	20.0
1.5	1.5	1.5	1.5	1.5	1.5	1.5	1.5	1.5	1.5	1.5	1.5
1.0	1.0	1.0	1.0	1.0	1.0	1.0	1.0	1.0	1.0	1.0	1.5
0.375	0.375	0.375	0.375	0.375	0.375	0.375	0.3	0.375	0.375	0.375	0.375
1.5	1.5	1.5	1.5	1.5	1.5	1.5	1.5	1.5	1.5	1.5	1.5
1.0	1.0	1.0	1.0	0.75	1.0	1.0	1.0	1.0	1.0	1.0	0.75
112.0	112.0	112.0	112.0	112.0	196.0	196.0	196.0	112.0	112.0	112.0	112.0
40.0	40.0	40.0	40.0	40.0	40.0	40.0	40.0	40.0	40.0	40.0	40.0
3.18	2.38	3.31									
.....	.....	1.44	5.38	0.38							
.....	.....	.....	.....	3.0							
.....	.....	.....	.....	.....	2.75						
.....	.....	.....	.....	.....	.....	3.12	5.18	5.18	6.18		
.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	3.18	3.18
0.55	1.13	0.74	0.65	0.67	0.83	0.86	0.83	2.13	1.11	1.23	0.98
0.18	0.35	0.28	0.42	0.33	0.08	0.20	0.56	0.75	0.54	0.17	0.49
0.78	2.20	1.42	1.67	0.85	0.55	0.51	0.99	0.73	0.52	0.81	0.49
0.033	0.055	0.045	0.057	0.016	0.04	0.070	0.098	0.078	0.086	0.086	0.064
0.006	0.007	0.07	0.005	0.009	0.003	0.003	0.004	0.002	0.003	0.003	0.008
2.55	1.19	2.58	2.58	2.17	1.69	1.79	1.66	2.30	0.44	2.26	3.7
			0.08	0.22 0.07							
2d	2d	2d	2d	2d	2d	2d	2d	2d	2d	2d	2d
1.0	1.0	1.0	1.0	½	¼	¼	¼	0.0	0.0	0.0	0.0
68.3	42.0	50.2	45.1	63.6	45.6	55.8	39.1	54.2	8.9	41.4	67.9
3.73	2.78	5.14	5.72	3.41	3.7	2.72	4.24	4.24	4.92	5.45	5.45

* 6 oz. nickel shot in charge.

The following numbered tests have been omitted from the table, for want of space: No. 94 and 101, conditions and results similar to those of No. 95 to 100. No. 103, similar to No. 102 and 104. No. 108, similar to No. 107 except in regard to nickel. No. 109 and 111, similar to No. 110. No. 112 and 114, similar to No. 113. No. 119, similar to No. 118 and 120.

tests with uranium metal. In experiments No. 94 to 97, inclusive, the ferro-uranium was broken to pea size; in the remaining tests, it was about ½-in. size.

On studying the results of these experiments, it may be noticed that in some tests there are several variable factors, so that conclusions as to what caused the results must be made with care. Due to the considerable mechanical loss of steel in pouring such small charges, the efficiency could not be calculated from the analysis of the steel and its weight. The uranium recovery is to a certain extent approximate. It is calculated on the assumption that all of the iron in the ferro-uranium combines with the weight of steel charged into the furnace.

This total weight is used as a basis for figuring the desired percentage

of uranium in the steel, which, compared with that actually secured, gives the uranium recovery. This method of computation is more accurate than one based on the weight of metal poured, in that it shows what may be expected under operating rather than experimental conditions.

The average uranium recovery in all of the 27 runs was 54.7 per cent.; when a steel containing less than 2 per cent. uranium was the object, the recovery was 67.8 per cent. Figures on average recovery under different conditions are given in Table 33.

TABLE 33.—*Average Recovery of Uranium under Different Conditions*

Average of all tests	54.7 per cent.
Steel containing less than 2 per cent. uranium	67.8 per cent.
Ferro-uranium added after skimming	70.7 per cent.
Ferro-uranium added before skimming	48.3 per cent.
Poured at once after adding ferro-uranium	60.2 per cent.
Poured ¼ min. after adding ferro-uranium	45.3 per cent.
Poured ½ to 1 min. after adding ferro-uranium	56.7 per cent.
Recovery with alloy No. 1 (25.1 % U)	65.1 per cent.
Recovery with alloy No. 3 (25.1 % U)	45.1 per cent.
Recovery with alloy No. 4 (24.85% U)	49.1 per cent.
Recovery with alloy No. 5 (29.2 % U)	45.7 per cent.
Recovery with alloy No. 6 (19.5 % U)	44.1 per cent.
Recovery with alloy No. 7 (37.55% U)	52.1 per cent.

All of the steels were cooled in lime. Some drilled with considerable difficulty, but no relation between ease of drilling and uranium content could be observed, because of cooling from and to different temperatures. All of the uranium steels are brittle when cooled slowly, and some when chilled. The slowly cooled steel has a coarse fracture, and the chilled a fine fracture. The shrinkage, with most of the steels, goes almost to the bottom of the ingot. This feature is not so marked on the high-carbon steels.

The following conclusions may be drawn regarding the addition of uranium to steel in quantities up to 4 per cent.

(1) Ferro-uranium[1] is a satisfactory agent for addition of uranium to steel. In comparison with uranium metal, the uranium probably stays in the steel better when added as ferro-uranium, because the iron acts as a carrying agent.

(2) A uranium recovery of at least 50 per cent. can be made with all grades of steel up to 4 per cent. uranium.

(3) A uranium recovery of 70 per cent. or better can be made with steel containing less than 2 per cent. uranium.

(4) The recovery is highest when the ferro-uranium employed contains from 25 to 37 per cent. uranium.

(5) When making additions of over 2 per cent. of uranium to steel, a ferro-uranium containing 50 per cent. uranium will probably be the best alloy to use. A 25-per cent. alloy means the addition of too much iron.

(6) A considerable proportion of the carbon and silicon in the ferro-uranium seems to enter the steel.

(7) For a steel containing less than 2 per cent. uranium, the ferro-uranium can be added in the ladle; but a higher percentage of uranium in the steel requires addition in the furnace, or chilling will occur.

(8) A higher uranium recovery is attained when the slag is skimmed, the ferro-uranium added, and the metal poured, than when the ferro-uranium is added without skimming, and the steel immediately poured. However, in the presence of a basic slag, a recovery of at least 50 per cent. of the uranium can be made.

(9) The results do not clearly indicate the proper length of time to leave the steel in the furnace after addition of ferro-uranium. Leaving it in for 1 min. before preparing to pour does not seem to diminish the uranium recovery to a great extent.

(10) The ferro-uranium should be added at about ½ in. size, without much fine material in it; this is especially advisable if added in the presence of slag.

(11) In general, it appears that ferro-uranium should be added after all the other ferro-alloys in use, and the steel should be poured as soon thereafter as possible. There are no data from the ferro-uranium experiments to show the effect of leaving the steel in the furnace for periods greater than 1 min.; but the results with uranium metal indicate that the recovery is highest when the steel is poured shortly after adding the uranium. A fair uranium recovery can be made with almost any type of slag, if the steel is not left in the furnace long enough to permit slagging of uranium.

Addition of Uranium to Steel in Crucible Furnace

Six heats of uranium steel were made at the plant of the Braeburn Steel Co., Braeburn, Pa., with the results shown in Table 34. The alloys had the following compositions: ferrotungsten, 71.75 per cent. tungsten; ferrochrome, 60.75 per cent. chromium; ferrovanadium, 38.52 per cent. vanadium. Ordinary 80-per cent. ferromanganese was used. The high-speed scrap was of varying analysis. The ferro-uraniums were as follows:

	No. 1 Per Cent.	No. 2 Per Cent.
Uranium	52.87	50.4
Carbon	4.74	4.95
Silicon	2.33	1.16
Vanadium	2.15	3.25
Iron	37.2	39.33

TABLE 34.—*Uranium Steel in the Crucible Furnace*

	121	122	123	124	125	126
Charge:						
Wash metal, lb	2.0	2.0	2.0	2.0	2.0	2.0
Ferrotungsten, lb	4.62	4.62	4.62	4.62	4.62	4.62
Iron, lb	50.0	50.0	50.0	50.0	50.0	50.0
Ferrochrome, lb	4.31	4.31	4.31	4.31	4.31	4.31
Ferrovanadium, lb	4.0	4.0	4.0	4.0	4.0	4.0
Ferromanganese, lb	1.25	1.25	1.25	1.25	1.25	1.25
High-speed scrap, lb	35.0	35.0	35.0	35.0	35.0	35
Ferro-uranium No. 1, lb	0.4	1.0	3.0	4.0	5.0	1.6
Ferro-uranium No. 2, lb						10.4
Analysis steel:						
Carbon, per cent	0.47	0.64	0.57	0.75	0.78	1.04
Manganese, per cent	nil	tr.	nil.	nil	nil	tr.
Silicon, per cent	0.16	0.32	0.14	0.13	0.16	0.36
Tungsten, per cent	7.50	7.57	8.53	8.37	8.15	6.76
Chromium, per cent	3.46	3.68	3.68	3.69	3.62	3.54
Phosphorus, per cent	0.024	0.032	0.023	0.023	0.020	0.024
Vanadium, per cent	1.38	1.67	1.69	1.87	1.81	2.13
Uranium, per cent	0.09	0.21	0.24	0.88	1.02	3.28
Desired per cent. uranium, 50-per cent. efficiency	0.10	0.25	0.75	1.00	1.25	3.00
Ferro-uranium in furnace, min	1.0	1.0	1.0	5.0	14.0	14.0

The results indicate that ferro-uranium can be added in the crucible as well as in the electric furnace, with about the same recovery of uranium in the steel.

Manufacture of Ferro-uranium

The materials used in the manufacture of ferro-uranium are uranium in the form of sodium uranate or uranium oxide, preferably the latter, fluorspar, petroleum coke, and steel turnings. All of these materials should be as free as possible from impurities, especially silica or silicon. The steel should not contain over 0.3 per cent. carbon, 0.15 per cent. silicon, 0.1 per cent. sulfur, 0.1 per cent. phosphorus. The uranium may be supplied also in the form of slag or residue from previous runs.

The most suitable furnace for the production of ferro-uranium is an electric furnace of the Siemens type, which has the crucible bottom as one electrode and a single suspended electrode. If it is not desired to keep the carbon content of the alloy below 4 per cent. the crucible of the furnace should be made completely of carbon or graphite rammed into place with pitch as a binder. When an effort is being made to keep the carbon as low as possible, the crucible should be lined entirely with ground dead-burned magnesite rammed into place with pitch as a binder, and the bottom electrode connection made by a water-cooled steel contact in the

magnesite. There should be a roof of silica brick over the crucible. The furnace may be of the stationary or the tilting type. Whatever type of electric furnace is used, it must be capable of intense concentration of heat, because of the high reduction temperature of uranium oxide, 1490° C., with a carbon reducing agent.

For making ferro-uranium, steel turnings, uranium oxide, petroleum coke, and fluorspar are mixed in the proper proportions and charged slowly at a rate dependent on the size of furnace and grade of alloy to be made; the mixture should not be charged so fast as to cool the furnace to any great degree. Regular interval of charging is essential. The metal is poured or tapped into an iron mold. In operation of the furnace, a portion of the charge collects on the wall, protecting the lining.

The reaction involved is that of simple carbon reduction, the reduced uranium alloying with the steel turnings to form ferro-uranium.

$$UO_2 + 2C = U + 2CO.$$
$$U_3O_8 + 8C = U_3 + 8CO.$$

Fluorspar is the only slag-forming material that can be used to obtain a reasonable recovery of uranium. It should be used in considerable quantity, and be of a very high grade.

A typical charge for the production of ferro-uranium containing 50 per cent. uranium is given below; also an analysis of the product. The uranium oxide contained 73.33 per cent. uranium, the coke 85 per cent. fixed carbon, the turnings 0.20 per cent. carbon.

Charge	Pounds	Product	Per Cent.
Steel turnings	15	Uranium	50.4
Uranium oxide	40	Carbon	4.95
Coke	7	Silicon	1.16
Fluorspar	15		

With a recovery of 75 per cent. of the uranium, 100 lb. of this charge will produce 57 lb. of ferro-uranium, and by resmelting the residue would produce (at 85 per cent. recovery of uranium) about 65 lb. of ferro-uranium.

The uranium recovery in the first smelting, using uranium oxide as a source of uranium, is from 60 to 75 per cent. of the uranium charged; in larger scale work, 85 per cent. might be recovered. On re-treating the residue, the total uranium recovery can be brought to from 85 to 95 per cent. The power consumption is 3 to 5 kw.-hr. per pound of ferro-uranium.

Characteristics of Ferro-uranium and Uranium Metal

Standard commercial ferro-uranium contains 35 to 50 per cent. uranium and 1.5 to 4 per cent. carbon. Ferro-uranium is brittle, has a high density, and when it contains over 20 per cent. uranium has a tend-

ency to be pyrophoric, this tendency increasing with the percentage of uranium. A 3 to 5 per cent. carbon alloy, with less than 15 per cent. uranium, has a dense non-crystalline structure, and is dull gray in color. When the uranium content reaches 18 to 20 per cent., small bright crystals appear; and as the alloy approaches 50 per cent. uranium, these crystals become larger, longer, and more distinct. From 18 per cent. up, the alloy has a brilliant, flaky fracture. At 18 per cent. it is hard to break, but at 50 per cent. uranium it breaks easily. With the commercial 50 per cent. alloy no graphitic carbon is present if the total carbon is under 5 per cent., but if there is over 5 per cent. carbon, this excess carbon is usually graphitic. Any of these alloys has an entirely different appearance if cooled from a white heat in water, this treatment giving it a granular and dirty appearance. If the carbon is less than 1.5 per cent., the alloy has a dense gray appearance and is difficult to break. Under the microscope, the uranium appears to be present as the double carbide of iron and uranium, $Fe_3C.U_2C_3$, in a eutectic mixture of iron and carbon.

Uranium metal with over 90 per cent. uranium has a dull black color. It is dense in structure, but powders easily. It is very pyrophoric, and draws a long spark when struck across steel or another piece of uranium metal. Uranium carbide is also pyrophoric, but is more flaky in structure than a mixture of uranium metal and uranium carbide.

Summary

The information contained in this paper has been presented in the hope that it will be of use to others during this period when the country needs all available information, and when there should be no secret process. No attempt is made to cover the manufacture of ferro-alloys beyond the metallurgical side. Furnaces and their operation are not discussed. The greater part of the data has been drawn from commercial operation, and in these cases the conclusions stated are brief. Where the information was gained by experiment, the development of the product through the research is covered in detail.

The metallurgical problems encountered in the manufacture of ferro-alloys have been the production of an alloy with a low percentage of carbon, a high percentage of the alloying element, and a low percentage of impurities, such as phosphorus and sulfur, that might be injurious to the steel. The alloy should have a melting point low enough to be used in a steel bath of ordinary temperature. To obtain this feature it has been necessary, in some cases, to produce an alloy with a high percentage of carbon and a comparatively low percentage of the alloying element, for example, ferrotitanium. The ferro-alloys with a low percentage of carbon are considerably more expensive than the high-carbon alloys, because of the extra refining necessary, which requires a large power consumption, and to a small extent increases the losses. In the case of

some alloys, for example, ferrovanadium and ferrotungsten, the steel manufacturer will not accept a high-carbon alloy, *i.e.*, containing over 1 per cent. carbon, under any conditions, because of the tendency to form carbides in the steel. It is also essential to produce a ferro-alloy of uniform composition, a difficult feature with some alloys. A ferro-alloy which is not fairly uniform is a source of trouble and expense in the manufacture of alloy steel, because it is not possible to regulate closely the composition of the steel.

Carbon control in the electric furnace depends upon two possible characteristics of the alloying metal: first, has it a strong affinity for carbon, or, reduced to fundamental principles, does it form carbides? Second, is the alloying metal easily oxidized? Some metals, such as chromium, manganese, uranium, and vanadium, form carbides even in the presence of a very small amount of carbon at the temperature of the electric furnace, resulting in a high-carbon alloy; while others—for example, silicon and tungsten—show a very slight tendency to form carbides. Some ferro-alloys, when refined to diminish the carbon, show a tendency to lose the alloy metal, because it oxidizes into the slag; examples of this are uranium and chromium. In the case of uranium, it is possible to lose practically all the uranium in the alloy in this manner, even when starting with a 60-per cent. ferro-uranium. On the other hand, tungsten does not oxidize, so that usually, when starting with a 70-per cent. tungsten, 3-per cent. carbon alloy, the refined product will show an increase of tungsten to 75 per cent. and a reduction of carbon to 1 per cent.

There are three general methods of producing low-carbon ferro-alloys: (1) The ore may be smelted with excess carbon to produce a high-carbon alloy with a low slag loss; after breaking up this high-carbon alloy, it may be charged with oxidizing slags to remove the carbon in a refining furnace. (2) The ore may be smelted with about the theoretical amount of carbon to produce a 1-per cent. carbon alloy with a high slag loss, but not followed by any refining process. (3) The ore may be smelted with a metal powder, such as silicon or aluminum, which reduces the metal to a product low in carbon, but high in silicon or aluminum, with a low slag loss; after breaking up this impure alloy, it can be charged with an oxidizing slag to remove the silicon or aluminum in a refining furnace. When using aluminum, it is not usually necessary to adopt a refining operation.

DISCUSSION

E. S. BARDWELL,* Great Falls, Mont. (written discussion†).—There are several points in connection with the manufacture of ferromanganese in the electric furnace which it seems to me might prove of interest in this connection. Recent observation of a number of electric furnaces

* Metallurgist, Anaconda Copper Mining Co. † Received Sept. 20, 1918.

varying considerably in size and transformer capacity, and all working under the same conditions, shows the small furnaces to be much more efficient in operation than the large furnaces.

The furnaces examined were all of the same type. The crucibles consisted of tanks varying in size from 7½ by 15 ft. by 4½ ft. deep, to 12 by 20 ft. by 9 ft. deep, with linings formed of carbon electrode butts. The smaller furnaces had transformer capacity rated at 1500 kv-a.; the largest furnace had a transformer rated at 3000 kv-a. All of the furnaces were operated on three-phase, 60-cycle current, the three electrodes being in a straight line. The voltages varied from 65 volts in the smaller to 100 volts in the larger furnaces. The charges consisted of manganese ore, coal, iron turnings, and limestone in proper proportions.

The small furnaces, operating at low voltage, were making good recoveries. The slag loss was low and the volatilization loss almost nil. The larger furnaces, operating at higher voltages, showed increased loss of manganese in the slag and greatly increased volatilization loss when working on the same charges as the smaller furnaces, or on charges similarly calculated.

The exact reasons for this state of affairs are difficult to state; a number of factors probably contribute to the results obtained in the large furnaces. First as to the effect of the higher voltage on manganese content of the slags; increasing the voltage of the furnaces apparently tends to promote irregular working. The current takes the path of least resistance; this means superheating in certain areas in the crucible and insufficiently high temperatures in adjoining areas. The partly reduced manganese in the cooler areas passes into the slag and escapes further reduction.

As to the effect of the higher voltages on volatilization losses, it is clear that the result of the excessively high local temperatures will be to augment volatilization. Furnaces operating at the higher voltages are also subject to the formation of accretions consisting of carbide and graphite, which tend to form with an excess of carbon. The formation of carbide robs the charge of a portion of its lime, rendering the slag composition very different from that calculated and desired. The electrodes are forced up, and volatilization is still further increased. The presence of carbide and graphite in the furnaces working at the higher voltages points to the existence of higher temperatures in these furnaces than in those operating at lower temperatures. The small furnaces rarely show signs of accretions, although both carbide and graphite are formed in them to a limited extent.

The advantages to be gained from having a large unit are obvious, although there is much to be said in favor of a number of smaller units. Assuming for the moment that a large furnace is desired, let us consider

the voltage to be employed. It is evident that the large furnace possesses no advantage over the small furnace unless we can get proportional or more than proportional production from it. Output, of course, is dependent on power input. If we represent the current by I, the voltage between phases by E, and the reactance and impedance by X and Z respectively, we may write the following equations:

$$I = \frac{E}{\sqrt{3}\,Z}, \quad Z = \sqrt{X^2 + R^2}, \quad \textit{Power input} = \sqrt{3}\,EI \times (\textit{power factor}).$$

Examining the first equation, we see that the current input may be increased either by increasing E or by decreasing Z. Experiments have shown that in any given furnace the reactance, X, is a constant for all except the very lightest loads; hence any decrease in the resistance, according to the second equation, will result in a decrease in the impedance and a corresponding increase in the current and power input.

The resistance of the charge may be decreased by substituting coke for a part or for all of the coal. If this is done and if the voltage is kept reasonably low, it should be possible to obtain increased power input and steadier operating conditions than are obtainable on the higher voltage furnace.

R. Korten[4] gives some interesting data regarding the melting of ferromanganese in the electric furnace, particularly in the Keller furnace. He states that melting without loss of manganese by volatilization is most certain if low voltage is employed, that the arc must be short, and the hearth offer as large a surface as possible. A uniform heating over a large surface and not an intense heating over a small surface should be sought.

While conditions are not exactly the same in electric furnaces for the smelting of manganese ores, my experience has been that better results are obtained where low voltages are employed and when the power input is properly proportioned to the size of furnace employed.

As to the subject of slags, when treating Montana carbonate ore containing 36.8 per cent. Mn and 6.79 per cent. SiO_2, using coal as reducing agent and marble as flux, it has been found possible to keep the slag loss below 3 per cent. of the manganese in the ore. The average slag analysis over a period of a week was as follows: Mn, 4.26; SiO_2, 34.37; Al_2O_3, 7.93; CaO, 44.83; MgO, 4.10; BaO, 0.88 per cent.

This result is extremely good when compared with the slag losses quoted by Mr. Keeney, which vary from 12 to 25 per cent. Results like these are easy to get in small furnaces working at low voltages. I feel confident that we shall be able to duplicate them in the large furnaces after sufficient study.

The greatest problem that we have to solve is the reduction of the volatilization losses. We know that it is possible to keep this loss low in the small furnaces, and doubtless a more thorough understanding of

[4] *Stahl und Eisen* (Mch. 14, 1912) **32**, 426.

furnace conditions will enable us to solve the problem in the case of the larger units.

H. W. GILLETT,* Washington, D. C. (written discussion†).—It would be very interesting if Mr. Keeney would give the phosphorus content of the ores and slags, as well as of the alloys, in Table 2 relating to ferromanganese, so that one could compute how much of the phosphorus in the ore went to the metal, how much to slag, and how much, if any, was lost by volatilization. In this connection it would be desirable to know also the dimensions of the furnace, especially the height of the column of charge.

Some recent experiments have indicated that, by deviating somewhat from established practice, it may be possible to eliminate a large proportion of the phosphorus from a high-phosphorus ore, and one would like to know just how much of the phosphorus is retained in the alloy in regular practice.

Has any feasible method actually been tested for the dephosphorization of a high-phosphorus ferromanganese?

On p. 56 the conclusion is drawn that for a ferro-uranium to contain any uranium the carbon must be over 2.5 per cent., and to hold any large amount of uranium, the carbon must be over 3.5 per cent. However, ferro-uranium has been made[5] containing 50 to 70 per cent. U and under 2 per cent. of carbon. This was produced by using an excess of UO_2 over carbon in the charge. Keeney's conclusion that the carbon content of the alloy cannot be controlled by the amount of carbon in the charge seems to be based on experiments with sodium uranate rather than with uranium oxide.

It will be noted in Table 32 that runs No. 97 and 98 show a recovery in steel respectively of 100 and 118 per cent. of the U in the ferro-uranium added. Runs No. 116, 117, 118, made under quite comparable conditions, show 54, 9 and 41 per cent. recovery.

No account of the method of sampling the steel, or of the method for chemical analysis of U in the steel, is given. The discrepancies referred to may possibly be explained by segregation of U, or by analytical error. Several workers had ill success with titration methods for the determination of U in steel, and in the preparation of 50-lb. billets of uranium steels of varying compositions, pronounced segregation of U has been found to be the rule rather than the exception, even when most earnest precautions were taken to prevent it. There is no question that it is difficult to cause a large proportion of the added U to be taken up by the steel, but quantitative figures on recovery may be greatly in error unless the absence of segregation of U, which might affect the sampling, is proved.

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* Chem., U. S. Bur. of Mines. † Received Sept. 27, 1918.

[5] H. W. Gillett and E. L. Mack: Preparation of Ferro-uranium. U. S. Bureau of Mines *Tech. Paper* 177 (1917) 27.

Electric-furnace Problems

BY J. L. MCK. YARDLEY,* PITTSBURGH, PA.

(New York Meeting, February, 1919)

THERE are two general classes of problems in connection with electric furnaces. First, those relating to the best utilization of the electrical power inside the furnace; second, those connected with the bringing of the electrical power to the point where it is to be utilized. Referring particularly to the latter class of problems, many calculations have been made, covering a wide variety of furnaces, to check the electrical capacity and power factor found from test. The following analysis, made for the purpose of determining the maximum capacity and approximate performance of a new furnace, to operate at 160 volts on a 60-cycle circuit, may be of interest.

The first characteristic necessary to determine is the inductance, since this is the factor that limits the amount of power which can be put into the furnace. Inductance is calculated from the dimensions of the furnace, according to formulas of the U. S. Bureau of Standards,[1] as follows.

Formula 99, for mutual inductance, is:

$$M = 2l\left[\log \frac{2l}{d} - 1 + \frac{d}{l}\right]$$

Formula 103, for self-inductance, is:

$$L = 2l\left[\log \frac{2l}{R} - 1 + \frac{R}{l}\right]$$

Both of these formulas are approximations which are nearly correct when l is great compared to d or R.

l = length of circuit, in centimeters,

d = separation between the conductors, in centimeters,

R = geometric mean radius of conductors. For a circular tube, R = the radius of the circle. For a solid circular conductor, $R = 0.7788 \times$ the radius of the conductor. For a rectangular conductor, $R = 0.2235 \times (a + b)$, a being the length, and b being the width of the rectangle.

To calculate the inductance of a three-phase circuit, it is customary to figure the inductance of one leg. In circuits where the three conductors

* Captain, Engineers, U. S. A.

[1] *Bull.* 8, U. S. Bureau of Standards (1912).

are arranged in a plane with equal spacing, the equivalent separation of all three conductors from each other is equal to $\sqrt[3]{2}$ times the actual separation. The inductance of one leg of the three-phase circuit is obtained by subtracting the mutual inductance M from the self-inductance L. This is because the action of the current returning in the other wires reduces the inductance of the wire that is being calculated.

The type of electric furnace under consideration is similar to that shown in Fig. 1. For purpose of calculation, the furnace circuit was divided into four parts: electrode, holder, flexible lead, and transformer

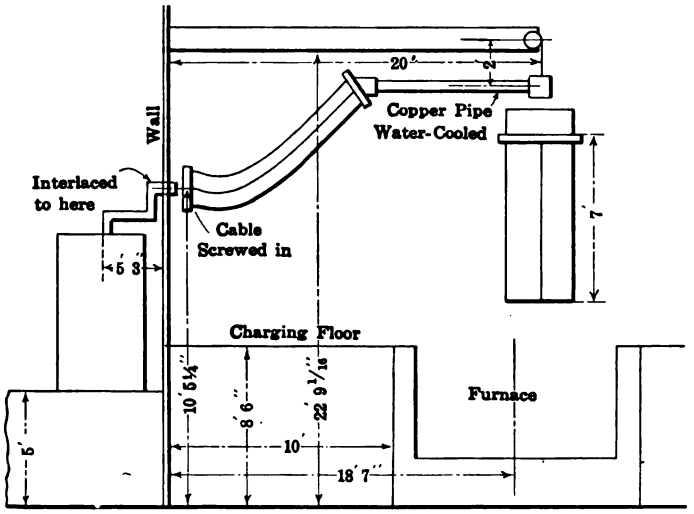

FIG. 1.—FURNACE *B*.

(including interlaced copper strap between transformer and flexible lead). Dimensions of these parts of the circuit are as follows:

	ELECTRODE	HOLDER	FLEXIBLE LEAD
l	345 cm.	282 cm.	305 cm.
d	192	192	192
R	40.9	7.62	18.15
L'	763 $(10)^{-9}$ henry	1453 $(10)^{-9}$ henry	1276 $(10)^{-9}$ henry

$L' = L - M$, and is the effective inductance of the conductor, taking into account both mutual and self-inductance, as previously outlined.

The reactance, in ohms, of any circuit may be calculated by the formula: $X = 2\pi fL'$ where f is the frequency and L' the inductance, as above.

Using this formula, the reactance of the electrode, holder, and flexible lead, on 60 cycles, is calculated as 0.001318 ohm. To allow for the reactive drop in the transformer and interlaced copper strap connections,

a reactive drop of 15 per cent. was assumed at full load. This means that with 30,000 amperes flowing into the furnace per phase, and 160 volts between phases, that is 92.3 volts between each phase and neutral, the reactive drop would be $0.15 \times 92.3 = 13.85$ volts. To obtain the reactance, this reactive drop in volts was divided by the current in amperes, giving 0.000462 ohm. Adding this to the reactance found for the electrode, holder, and flexible lead, a total reactance for the furnace circuit of 0.00178 ohm was obtained.

As such calculations are necessarily but approximations, it is important to be able to check by tests on similar furnaces. In this case, tests on two other similar furnaces were fortunately available. These were furnace *A* which operates at 158 volts, 25 cycles, and furnace *B* which

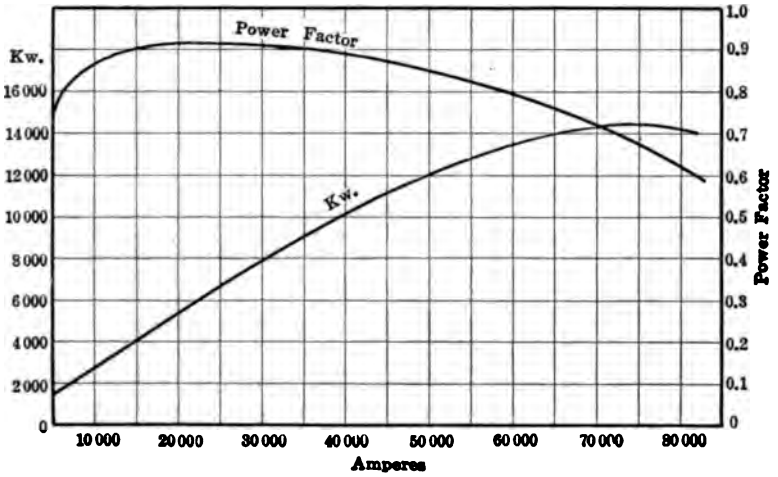

FIG. 2.—PERFORMANCE OF FURNACE *A*. 25 CYCLES.
(Transformer exciting current equivalent to 4620 low-tension amperes.)

operates at 97 volts, 60 cycles. The performance curves for these two furnaces are given in Figs. 2 and 3. It will be noted that furnace *A* operates at practically the same voltage as the proposed furnace, though at a different frequency, while furnace *B* operates at the same frequency though at a different voltage. By analysis of the test curves, Figs. 2 and 3, the reactance of the furnaces was determined as 0.00205 ohm for furnace *B* and 0.000741 ohm for furnace *A*. Correcting the value of reactance for furnace *A* to 60 cycles, it is 0.001775 ohm. It happened, however, that the overall power factor shown on the curve for furnace *A* included the transformer exciting current, which was in the neighborhood of 15 per cent. of the full-load current value. Such an exciting current is abnormally high for furnace transformers, and in the layout for the proposed furnace it was planned to use transformers having so small an exciting current as to have a negligible effect upon the power factor.

Taking the power factor shown on the curve for furnace *A* at 60,000 amperes (which current corresponds at 25 cycles to 25,000 amperes at 60 cycles) as being due to reactance only and no exciting current, the reactance of furnace *A* would be 0.000934 ohm; correcting this for 60 cycles, the value is 0.00224 ohm.

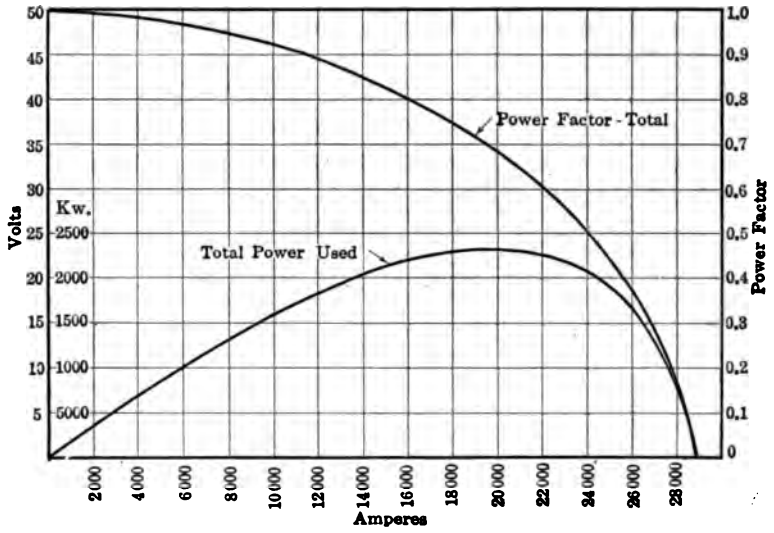

FIG. 3.—PERFORMANCE OF FURNACE *B*. 60 CYCLES.

In comparing the three furnaces, their dimensions were considered with regard to their effect on inductance, remembering that inductance varies directly with the length of the circuit and with the logarithm of the ratio of separation to radius of conductor. The dimensions of the furnaces, as far as available, are as follows:

	Proposed Furnace	Furnace *A*	Furnace *B*
Electrode length	11′ 4″	10′	
Electrode section	18″ × 54″	15″ × 60″	
Electrode separation	5′ 0″	7′ 6″	4′ 9″
Electrode holder length	9′ 3″	11′ 3″	
Electrode holder section	6″ dia. tube	7¼″ × 10″	
Flexible lead length	10′	11′ 6″ to 19′ 6″	
Reactance at 60 cycles	0.00178 ohm (calculated)	0.00224 ohm (test, modified)	0.00205 (test, modified)
Total length of one leg	30′ 7″	.....	29′ 2″

It will be noticed that the calculated reactance of the proposed furnace is less than the total reactance of the other two. The electrode separation on the proposed furnace is less than on furnace *A*, and the flexible

leads are shorter, so that the inductance should be less. The total length of circuit for the proposed furnace is slightly greater than that of furnace *B*, and the separation is also greater; so that, provided the cross-section of the circuit is the same, the proposed furnace should have higher inductance than furnace *B*. Since no data were available on the section of the conductors used for furnace *B*, it was felt that 0.0021 ohm would be a conservative figure to employ as the value of the reactance of the proposed furnace, taking into account the test and data on furnace *A*.

The performance of the proposed furnace, calculated with the reactance of 0.0021 ohm, is shown by the curve in Fig. 4, compared with the test curves of the other two furnaces modified to a uniform basis of 160 volts, 60 cycles. The tests on furnace *A* were changed to correspond to

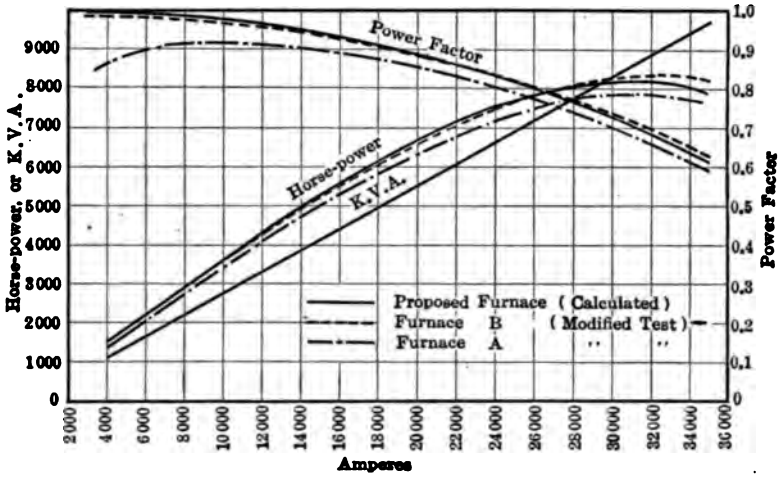

FIG. 4.—COMPARATIVE PERFORMANCE OF 160-VOLT, 60-CYCLE FURNACES.

the proposed conditions by reducing the current in the ratio of 60 to 25. This should give a fair comparison of power factor for the reason that, since the voltages are the same in both cases, the reactance voltage drop would be the same for the same power factor. Since the reactance voltage drop is proportional to the product of current and frequency, if the current is changed in the inverse ratio to the frequencies, the reactance voltage should remain the same. The test results on furnace *A*, modified in this manner, are plotted in Fig. 4, with the calculated performance of the proposed furnace.

In order that the test results on furnace *B* might be on a comparative basis, they were treated as follows: Since the operating voltage of furnace *B* is 97 while that of the proposed furnace is assumed as 160, a given power factor would allow only (97 ÷ 160) times as much reactance voltage drop in furnace *B* as for the proposed furnace. Certain current

values were chosen on the curve for furnace *B* and the corresponding power factors were tabulated. The reactance factor corresponding to a given power factor is, of course, the sine of the angle whose cosine is the power factor. Multiplying the reactance factor by (97 ÷ 1.73) gave the reactance voltage drop per leg. This reactance voltage drop was then multiplied by the ratio of the voltage (160 ÷ 97) in order to obtain the corresponding reactance voltage at the same power factor and at 160 volts. Since this reactance voltage is equal to the product of the current and the reactance, and since the reactance of the proposed furnace is approximately equal to the reactance of the furnace *B*, the ratio of the current which will flow into the new furnace to the current flowing into furnace *B* is the same as the ratio of the reactive voltages, which again is the same as the ratio of the voltages, that is 160 to 97. The curve for furnace *B* modified for 160 volts is shown in Fig. 4 together with the curves for furnace *A* and the new furnace.

The above is simply an example of the way in which heavy-current furnace problems may be worked out. The important thing is to have plenty of test data to refer to for check purposes. As shown by Fig. 4, the maximum useful input which the furnace under consideration could possibly take would be in the neighborhood of 31,000 amperes at 160 volts, approximately 71 per cent. power factor, three phase, which would be 8600 kv-a. or approximately 6000 kilowatts.

Acknowledgment is gladly made to R. H. Willard, who, under the writer's direction, made the detailed analysis of test results and the calculations for this furnace.

DISCUSSION

Jesse L. Jones,* Pittsburgh, Pa. (written discussion†).—From a commercial standpoint, the most important characteristic of the electric furnace is the ease with which it lends itself to intensive production. The melting of non-ferrous metals, as usually done in crucible, open-flame, reverberatory, and similar furnaces, has certain disadvantages which are absent in the electric furnace, viz.:

1. There is contact of the material being melted, for a long period, with air, which results in the formation of oxides and dross and, in the case of some metals, nitrides. If clean castings are desired from such metal, skim gates, risers, and similar devices must be used. The large amount of extra metal required greatly increases costs while the dross, if once formed, can never be perfectly removed, so that absolutely clean castings cannot be obtained. Moreover, any dross included in a casting, being of an abrasive nature, adds to the expense of the machining.

2. A prolonged period of melting will cause absorption of hydrogen,

* Westinghouse Elec. & Mfg. Co. † Received Jan. 17, 1919.

carbon monoxide, and other gases from the products of combustion of the fuel. The release of these gases during solidification results in spongy castings.

3. Contact of molten metals for long periods with the refractory materials of crucibles, furnace linings, etc. results in silicon, silicates, iron, iron oxides, alumina, etc. being alloyed, or entangled, with the molten metal and resulting marked disadvantages.

4. Slow melting largely increases volatilization losses in the case of many metals. It requires more floor space for molds, causes loss of time by employees if the molds are ready in advance of the metal, and, where crucibles are used, diminishes the life of the crucible.

Possibly greater speed of melting has been obtained in the oil-fired open-flame furnace, up to the present, than in any other type of furnace. A record has been made of 26 heats of 500 lb. each of brass in a working day, in a furnace of this type. Individual heats of 500 lb. have been made in 15 min. actual melting time, and heats of 300 lb. in 8 min. This rapid work has been made possible by the liberal use of fuel and the stored heat in the furnace linings.

Mr. Yardley has shown that a very large input of current may be made to the electric furnace and it is the object of this discussion to emphasize the economic advantages that may result from the satisfactory application of such heavy currents. Possibly we may never be able to melt metals in the electric furnace instantaneously, in a reducing atmosphere and with a minimum of volatilization loss, but the electric furnace has more possibilities of approximating these ideal melting conditions than any melting device of which we have any present knowledge.

J. L. McK. YARDLEY (author's reply to discussion*).—This is about the maximum amount of power that can be brought to a single electric furnace where the supply frequency is 60 cycles and only one bank of furnace transformers is employed. Where two banks of furnace transformers are employed in parallel electrically, supplying power to the same electric furnace, and the impedance in the furnace circuit is correspondingly reduced, a much greater amount of power can be used. Also, a much larger quantity of power can be used in one furnace where the supply frequency is only 25 cycles. Thus, with a supply frequency of 25 cycles there are individual electric furnaces actually taking about 17000 kv.-a. By using two banks of transformers it should be possible to do approximately the same thing where the supply circuit is 60 cycles.

* Received Apr. 18, 1919.

Basic Refractories for the Open Hearth

BY J. SPOTTS MCDOWELL* AND RAYMOND M. HOWE,† PITTSBURGH, PA.

(New York Meeting, February, 1919)

Preparation and Use.—Magnesite is an important refractory in open-hearth, heating, and electric furnaces for steel-making and in many of those employed in the metallurgy of copper and lead. It is sold in the form of brick, finely ground "furnace magnesite" for bricklaying, and dead-burned grains for making and repairing furnace bottoms. The latter are a mixture of granules varying in size from pieces of about ⅝ in. (1.6 cm.) diameter to very fine but sandy particles. Dead-burned magnesite results from calcining the crude or lightly burned mineral at a temperature that will not merely drive off practically all the CO_2, but will also cause sintering of the particles. During this process the pieces shrink considerably and become hard, dense, and inert to atmospheric moisture and CO_2; under-burned material, on the other hand, will hydrate on exposure to the air. A small percentage of ferric oxide seems to be necessary for the production of a satisfactory sinter; from 4.5 to 8 per cent. in the dead-burned grains is considered the most desirable amount.

Dolomite has been little used for brickmaking in the United States, but it is prepared for use in the granular condition calcined or "double-burned" and is the principal ingredient of several materials offered for sale under various trade names for refractory purposes. Dolomitic refractories are almost wholly confined to the open-hearth and electric furnaces, where they are used for fettling and as substitutes for magnesite.

Much more magnesite and dolomite are used for basic open-hearth steel-making than for all other refractory purposes. The hearth of the furnace is usually built up of magnesite brick and dead-burned grain magnesite so laid that the brick base is protected by a working bottom of the granular material. The latter is sintered into place in layers ½ to 1 in. (1.25 to 2.5 cm.) thick to a total depth of 12 to 18 in. (30 to 45 cm.) at the center of the furnace. After each heat, burned dolomite is thrown against the banks as high as it will stick, and all holes in the bottom are filled. At most plants such holes, at the end of each week, are also carefully filled with dead-burned grain magnesite. For temporary

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†Fellow, Mellon Institute.

patching dolomite is generally used as it sets more quickly than magnesite and its first cost is less.

Prior to 1914, the world's supply of refractory magnesite came almost wholly from the crystalline deposits of Austria-Hungary. Its superiority lies in its high refractoriness and long range of vitrification, which enable it to frit together at high temperatures without fusion or excessive softening. These properties are imparted by a fairly high content of iron oxide, together with an extremely low percentage of harmful impurities. The dead-burned Austrian magnesite sold in the United States had the following range of analysis: MgO, 83.7 to 87.3 per cent.; CaO, 1.9 to 3.9 per cent.; SiO_2, 1.1 to 4.1 per cent.; Fe_2O_3 and Al_2O_3, 4.7 to 8.6 per cent. With the cessation of shipping from Austria, magnesite was imported from Greece and Canada, domestic deposits were developed, and as a temporary expedient dolomite was substituted for magnesite where practicable. The deposits of this country are now developed sufficiently so that an adequate supply is available and ample reserves are blocked out to last many years. Table 1 shows the great expansion of domestic production since 1914. All figures are reduced to a calcined basis, assuming 2 tons of crude magnesite equal to 1 ton of calcined.

TABLE 1.—*Total Magnesite Consumption of United States on a Calcined Basis*

	1913	1914	1915	1916	1917
Imported, net tons	173,719	128,494	51,458	46,941	19,093
Domestic, net tons	4,816	5,646	15,250	77,487	158,419
Total, tons	178,535	134,140	66,708	124,428	177,512

At the beginning of the war, the refractory manufacturers were crippled by a lack of calcining facilities, since the Austrian material had been imported in the calcined condition. The low iron content of the available magnesite caused additional difficulties. In order to confer the proper sintering and bonding properties, it became necessary to add iron oxide and to incorporate it thoroughly by burning at an extremely high temperature. Magnesite thus treated, if sufficiently low in harmful impurities, is a high-grade refractory and has given conspicuously satisfactory service.

Grecian Deposits.—These deposits are of the amorphous type and had been operated on a large scale for many years but their output was rapidly increased after the war began. Considerable magnesite was imported from Greece in 1915 and 1916, but shipments were suspended late in the latter year. Much of it is of exceptional purity, as shown by

the following typical analyses of crude rock; though some of the imported material was very high in CaO and SiO_2, due doubtless to improper selection at the mines. The percentage of impurities after calcination will be practically double these figures on account of the loss of CO_2.

ANALYSES OF CRUDE MAGNESITE FROM GREECE

	Per Cent.	Per Cent.	Per Cent.	Per Cent.	Per Cent.
SiO_2	0.67	0.46	1.19	1.63	2 28
Al_2O_3 and Fe_2O_3	0.30	0.50	0.43	1.36	0.19
CaO	0.92	1.24	0.80	1.44	1.56
MgO	46.06	46.22	45.83	45.75	44.78
Ignition loss	52.16	51.51	50.42	49.88	49.76
	100.11	99.93	98.67	100.06	98.57

Magnesite of Grenville District, Quebec.—This magnesite is white to grayish, finely crystalline, and high in lime due to dolomitic inclusions. Wilson,[1] in 1917, estimated that there were in sight a little under 700,000 tons in the deposits of the district containing less than 12 per cent. lime (equivalent to over .20 per cent. lime when burned). It has been stated that when calcined and mixed with furnace slag or dead-burned with iron ore, the Canadian magnesite has given satisfactory service in furnace bottoms.[2] However, the leading refractory manufacturers prefer the purer mineral of California and Washington, and endeavor to maintain the CaO content below 4 per cent. in the dead-burned material. During the period of greatest scarcity of magnesite, the Canadian product was utilized by several companies in the manufacture of magnesia brick as a minor constituent of the mix, but this practice has probably been entirely abandoned. It is, however, being employed for the production of dead-burned grain magnesite, by sintering with iron oxide in rotary kilns.

California Magnesite.—This magnesite, which is of the amorphous type, is found at many places. It occurs in the form of veins, lenses, and stockwork in serpentine; the deposits are usually small. The largest and most important are those near Porterville, Tulare Co., and St. Helena, Napa Co. Prior to 1914, about 10,000 tons were being produced annually, hence the mines had not become sufficiently developed or equipped to produce large tonnages before the summer of 1916. Transportation was a serious problem in the rainy seasons, as nearly all of the deposits are a number of miles from the railroad and hauling to the cars is done by means of trucks. Variations in quality at first caused the users considerable difficulty. It was often hard to keep the silica and lime contents within the proper limits, due to the impurity of some deposits and the

[1] M. E. Wilson: Magnesite Deposits of Grenville District, Quebec. Canad. Dept. of Mines *Memoir* 98 (1917) 52.

[2] A Stansfield: *Iron and Coal Tr. Rev.* (Jan. 12, 1917) **94**, 31.

lack of uniformity in others, as well as to the common inexperience of the operators. Most of these difficulties were gradually lessened and in 1917 the production had risen to 211,663 net tons of crude magnesite.[3] In the meantime, a considerable number of calcining plants had been erected near the mines, in order to give the rock a light burning before shipment. Since a loss in weight of about 50 per cent. occurs in this burning, a material saving in freight resulted. Recently the California production has fallen off, due to the competition of the larger and less expensively operated deposits of Washington. The first analysis, which

TYPICAL ANALYSES OF CALIFORNIAN CRUDE MAGNESITE

	Per Cent.	Per Cent.	Per Cent.	Per Cent.	Per Cent.
SiO_2	3.25	5.18	1.10	3.86	1.55
Fe_2O_3 and Al_2O_3	2.20	1.10	0.40	0.80	0.45
CaO	1.25	3.34	Trace	2.04	1.38
MgO	43.87	41.92	46.54	43.47	45.68
CO_2	49.53	48.78	51.20	49.48	50.97
	100.10	100.32	99.24	99.65	100.03

is uncommonly high in iron for magnesite of this type, is representative of a unique deposit near St. Helena, Napa County.

Magnesite Deposits of Washington.—These deposits have become a most important factor in the market since their discovery in 1916; 715 tons of crude rock were mined and shipped in the latter part of that year, and 105,175 tons were produced in 1917. The mineral is finely to coarsely crystalline and shows many variations in color from white to gray, pink, red, and black. Mining is done by both open-quarry and underground methods. It is reported that diamond drilling at the Finch Quarry of the Northwest Magnesite Co. has proved the existence of more than 1,000,000 tons, and that on more than one of the properties an estimate of 1,000,000 tons within 200 ft. of the surface is reasonable.[4] Prior to the erection of calcining plants at the properties many thousand tons of crude rock had been shipped in which the silica content was uniformly below 3.5 per cent. and lime less than 1.5 per cent. The analyses given below are typical.

TYPICAL ANALYSES OF WASHINGTON MAGNESITE

	Per Cent.	Per Cent.	Per Cent.	Per Cent.
SiO_2	1.9	3.3	4.3	0.6
Fe_2O_3 and Al_2O_3	1.0	1.0	0.8	1.2
CaO	1.7	1.4	1.1	0.5
MgO	45.2	44.8	45.0	46.4
Co_2	49.7	49.3	49.5	51.0
	99.5	99.8	100.7	99.7

[3] C. G. Yale and R. W. Stone: Magnesite in 1917. *Mineral Resources of the United States*, 1917 (1918) Pt. 2, 65.

[4] Yale and Stone: *Op. cit.*

Calcination is now being done near the quarries, and dead-burned magnesite for refractory use to which iron oxide is added during manufacture is being produced in rotary kilns. The material as marketed by one producer has the following average analysis: MgO, 82.5 per cent.; CaO, 3.4 per cent.; SiO_2, 6.5 per cent.; Fe_2O_3 and Al_2O_3, 7.25 per cent.; loss on ignition, 0.5 per cent.

Dolomitic Refractories.—For many years prior to the war few basic bottoms in America had been built of dolomite, which was little used except for temporary patching and the fettling between heats. Since 1914, however, calcined dolomite and specially prepared dolomitic materials have been common substitutes for magnesite, being placed in the furnace in the same manner as the grain magnesite. Within recent months the tendency on the part of steel makers has been to displace burned dolomite by specially prepared dolomites for the temporary patching and to return to the use of high-grade magnesite for all original installations and major repairs.

In the calcination of dolomite either vertical or rotary kilns may be used. In the latter case the rock is usually crushed to pass approximately a ⅝-in. (16-mm.) screen, the fine powder screened out, and the granules burned. It is usually held that for best results the heat applied must be sufficiently intense not merely to drive off the CO_2, but to cause the pieces to shrink and become hard and dense. The rock is frequently "double-burned" by being heated in one kiln merely high enough to drive off most of the CO_2, and further shrunk in a second at a higher temperature. The usual range of composition of the crude rock is CaO, 28 to 35 per cent.; MgO, 14 to 20 per cent.; SiO_2, 1 to 7 per cent.; Al_2O_3, 0 to 4 per cent.; Fe_2O_3, 0.5 to 5 per cent.; CO_2, 43 to 46 per cent. As high a magnesia content as possible is considered desirable.

Burned dolomite as ordinarily prepared air-slakes readily on account of the high percentage of lime. For that reason it cannot be kept in stock for any length of time without deterioration and should be made just before using. A dolomite bottom is likely to show considerable slaking after a shut-down of even comparatively brief duration. Since dolomite does not set as solidly as magnesite, patches of a dolomite bottom are more apt to become detached and float in the bath.

Special Dolomitic Refractories.—Within the last few years, numerous investigations have been undertaken in the attempt to overcome the marked slaking properties and other defects of calcined dolomite. As a result several articles consisting of specially prepared dolomite have been placed on the market under various trade names, some of which are superior to ordinary calcined dolomite, being more resistant to atmospheric slaking and giving better service in the furnace. In most of these preparations the granules are coated or impregnated with pulverized basic slag, iron ore, or similar material by burning at a high tem-

perature in a rotary kiln. The attempt has likewise been made to coat the crushed rock with a pulverized slag suspension in water, so that upon burning a protective coating will be formed on the surface.[5] In other cases the slaking properties are diminished by using an impure dolomite, preferably one high in iron. The analysis considered most desirable may be obtained by mixing raw rock of different compositions.

One patented material is prepared by grinding together dolomite and iron ore to form an intimate mixture and calcining at a temperature reported to exceed 2850° F. (1565° C.). The finished product for furnace bottoms is said to show the following range of analysis: CaO, 42 to 55 per cent.; MgO, 38 to 25 per cent.; SiO_2, 6 to 13 per cent.; Fe_2O_3 and Al_2O_3, 8 to 14 per cent.

Another compound consists of an intimate mixture of granular dolomite and about 10 per cent. basic open-hearth slag. The manufacturers prefer hard, compact and dense varieties of dolomite, as high as possible in magnesia. It is said that the granules, after calcination in a rotary kiln at about 2800° F. (1540° C.), are mixed, upon coming from the kiln and while still hot, with the granulated slag and the mixture is sent through a rotary cooler. It is claimed by the manufacturers that material thus prepared may be transported without slaking when the magnesia content is high and the dolomite is extremely hard burned. The average analyses of several shipments is given herewith: CaO, 48 to 55 per cent.; MgO, 29 to 32 per cent.; SiO_2, 4 to 11.5 per cent.; Al_2O_3, 3 to 6.5 per cent.; Fe_2O_3, 4 to 7 per cent.

Laboratory Tests

The relative value of different materials used in the open-hearth furnace is measured by the service obtained and by that alone. Neither theoretical considerations nor the results of laboratory tests can be expected to demonstrate precisely what these relative values may be, although much can be learned from such studies. Furnace conditions cannot be reproduced in laboratory tests; these can merely show the relative resistance of different materials to destructive forces logically assumed to be similar to certain of those in the furnace. With these facts in mind, the authors some months ago planned a brief series of tests designed to compare properties of various basic materials that evidently affect the durability of these materials in the furnace. The results of these experiments, which were carried out under the direction of Raymond M. Howe at the Mellon Institute of Industrial Research of the University of Pittsburgh, are described here.

[5] A. V. Bleininger: Some Aspects of the Testing of Refractories. *Proc.* Eng. Soc. Westn. Pa. (1916–1917) **32**, 612.

The properties of the basic refractories which evidently affect their value for furnace use are

1. Resistance to slaking.
2. Resistance to erosion or scouring action of the bath.
3. Resistance to the chemical action of impurities in the bath.

The wearing away of the basic bottom is due to the abrasive action of the bath, particularly during the frothing accompanying the process of oxidation, the scouring out of holes caused by boiling, chemical reactions between constituents of the hearth and of the bath or slag, and possibly a few other minor causes.

It is well known that all basic refractories soften at high temperatures and it is believed that the softer the bottom the more rapidly will the scouring action take place. Other things being equal, the material that shows the least degree of softening at operating temperatures should give the longest service.

Corrosion of the lining due to chemical action is harmful, not only because of the wearing away of the refractory but because the surface of the hearth becomes weakened and less resistant to scouring. An investigation regarding the relative resistance of various basic refractory products to the action of corrosive melts is included in the tests here described.

Basic Materials Studied.—The six commercial products studied include two magnesites, two dolomites, and two specially treated dolomites. Throughout the remainder of the paper these will be designated by the letters *A*, *B*, *C*, *D*, *E*, and *F*. The material designated as *A* was prepared from Washington magnesite, ground with ferric oxide, and dead burned. *B* is a similar product prepared from Canadian magnesite. *C* is a preparation made by coating dolomite granules at a high temperature with basic open-hearth slag. *D* is made by mixing together iron oxide and dolomite and calcining at a high temperature. *E* is a relatively impure dolomite that has been calcined. *F* is a rather pure calcined dolomite. The analyses of these materials at the time of arrival are given in

Table 2.—*Analyses of Basic Materials*

	A	*B*	*C*	*D*	*E*	*F*
Moisture and ignition loss per cent	0.43	0.33	0.47	1.00	3.70	9.90
Silica, per cent	7.36	4.56	11.46	9.84	6.68	2.16
Alumina, per cent	2.33	2.08	3.91	1.86	3.89	0.90
Ferric oxide, per cent	4.81	7.88	4.41	4.54	1.69	1.10
Lime, per cent	3.12	18.94	47.92	48.70	49.84	51.00
Magnesia, per cent	81.60	66.40	32.10	34.73	34.50	34.85
Total	99.63	100 19	100.27	100.67	100.30	99.91

TABLE 3.—*Recalculated Analyses of Basic Materials*

	A	*B*	*C*	*D*	*E*	*F*
Silica, per cent	7.40	4.57	11.52	9.94	6.93	2.40
Alumina, per cent	2.34	2.08	3.93	1.88	4.03	1.00
Ferric oxide, per cent	4.83	7.89	4.43	4.58	1.73	1.22
Lime, per cent	3.13	18.96	48.15	49.19	51.64	56.61
Magnesia, per cent	81.93	66.66	32.26	35.18	35.75	38.69
Total	99.63	100.16	100.29	100.77	100.08	99.92

Table 2 and these analyses recalculated on a zero ignition loss basis to make them more comparable in Table 3.

Slaking Tests.—From the very manner in which basic materials are prepared, it can be assumed that a low ignition loss and the least possible tendency to slake are essential properties. It is said that these materials after slaking "powder" more or less when they give off CO_2 or water during heating. A part of this powder is carried out of the furnace by the draft, resulting in an additional loss in material, as well as injury to the checkers. It is suggested that the driving off of moisture and CO_2 under the action of the heat may also interfere with proper setting of the granules to a dense mass and thereby decrease the resistance to erosion.

Two methods of procedure were followed in the slaking tests. In the first, the materials were allowed to stand exposed to the air and each month a sample was taken, dried at 110° C., and its ignition loss determined. The results of this series of tests are not yet available. The second procedure consisted of moistening the different materials from time to time with water. Samples were then taken at intervals of 5 days, dried at 110° C., and the loss on ignition determined. The results are given in Table 4. Table 5 represents the ignition loss of recalcined *E* and *F*; it was deemed necessary to recalcine these two products because of their high loss at the beginning of the test.

TABLE 4.—*Percentage Loss on Ignition at Regular Intervals*

	A	*B*	*C*	*D*	*E*	*F*
Loss at beginning of slaking test	0.02	0.20	0.08	0.95	3.93	14.04
Loss after 5 days	0.48	0.69	1.20	5.67	18.45	25.52
Loss after 10 days	0.95	1.52	1.62	8.35	20.00	26.30
Loss after 15 days	1.24	1.94	2.39	10.10	23.30	33.00
Loss after 20 days	1.36	1.92	3.01	10.92	23.66	29.70
Loss after 25 days	2.09	2.65	6.30	12.00	24.33	25.99
Loss after 30 days	2.54	3.25	4.63	14.99	24.67	30.81

TABLE 5.—*Percentage of Loss Upon Ignition of Recalcined E and F*

	E	F		E	F
Loss after recalcination.	1.55	1.29			
Loss after 1 day.....	5.16	9.74	Loss after 10 days....	15.80	
Loss after 2 days.....	9.40		Loss after 12 days.....		23.53
Loss after 3 days.....	12.79	15.81	Loss after 15 days.....	17.09	
Loss after 5 days.....	12.81	16.43	Loss after 25 days	17.36	

The results given in these tables are also plotted graphically in Fig. 1. Table 6 gives the analytical and slaking data.

TABLE 6.—*Analytical and Slaking Data*

Material	A	B	C	D	E	F
Order of resistance to slaking....	1st	2d	3d	4th	5th	6th
Lime, per cent.....	3.16	18.96	48.15	49.19	51.64	56.61
Magnesia, per cent.....	81.72	66.66	32.26	35.18	35.75	38.69
Silica, alumina, and ferric oxide, per cent.....	14.70	14.54	19.88	16.40	12.69	4.62

It is to be observed that in the case of both magnesites and dolomites the slaking tendency increases with the lime content and that in the case of the dolomites it varies inversely with the sums of the iron oxide, silica, and alumina contents. The state of these impurities, whether they are natural or artificial, undoubtedly exerts an influence that these tests do not reveal. In the latter case the thoroughness with which the granules are coated or impregnated with ore or slag is an important factor. The magnesite low in lime, *A*, showed a greater resistance to slaking than the one high in lime, *B*. The low-slaking tendency of *C* apparently proves that the granules were well protected by the slag coating. It is somewhat higher in impurities and lower in lime than *D*, a treated dolomite that slaked badly. *E* was the more stable of the two calcined dolomites; its condition at the time of receipt was far superior to that of *F*. It is believed that these two cases serve as the best examples of the effect of lime content and fluxes upon the rate of slaking. The air-slaking series, although incomplete, indicates that the different materials stand in the same relative order as given above.

A study of the figures given indicates that the lime content should preferably be as low as possible and that iron oxide, alumina, and silica, within limits dependent on the character of the dolomite and its treatment, are necessary in dolomitic refractories, if the slaking is to be kept within practicable limits; further, that a dead-burned magnesite free from lime, and possibly containing a fairly high percentage of impurities (10 to

15 per cent. of SiO_2, Al_2O_3, and Fe_2O_3) is the most suitable basic refractory for withstanding the conditions of this test. These conditions represent the influences to which basic materials may often be subjected in cars or when stored in the plant. Pure calcined limestone is the poorest material to withstand such conditions. Other combinations of the lime-magnesia series will arrange themselves in intermediate positions according to the lime, alumina, silica, and ferric-oxide content, and the temperature of burning.

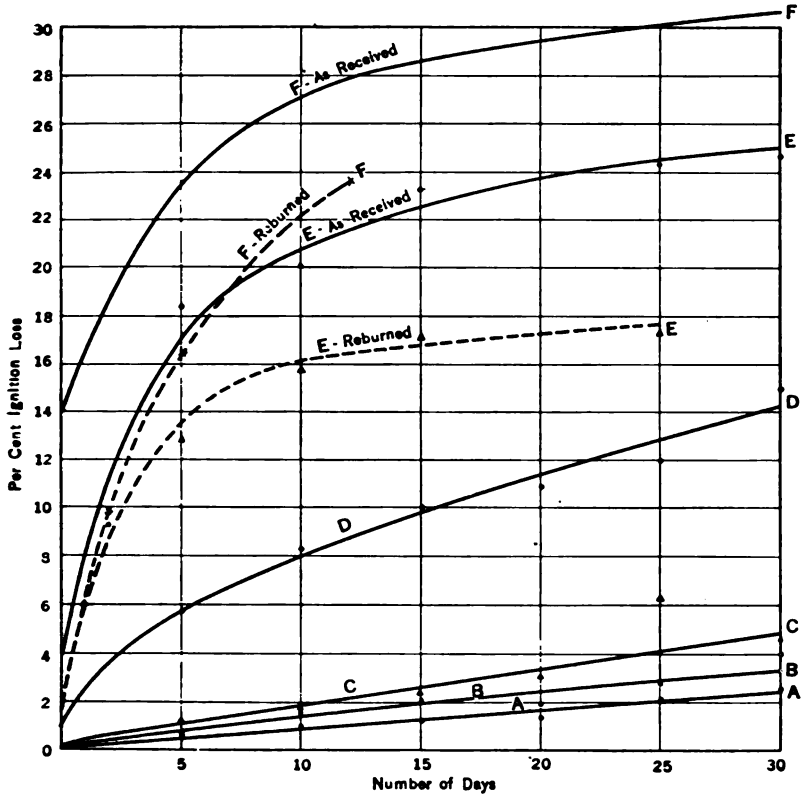

FIG. 1.—INCREASE OF LOSS ON IGNITION OF BASIC MATERIALS UNDER THE INFLUENCE OF MOISTURE.

Action Between Basic Materials and Fireclay or Silica Brick.—Failure is liable to result because of chemical reactions when basic materials are in contact with fireclay or silica brick at high temperatures. The latter are not uncommonly used in contact with magnesia brick above the slag line of the open hearth, but this is not considered the best practice and a buffer layer of chrome brick is desirable. With this exception the basic materials are generally carefully separated from silica and clay brick in furnace practice. In view of these facts the following tests were made to

show the relative limits of safety when the materials mentioned are heated in contact. Moreover, it was believed that the behavior of the basic products in this test would afford an indication of their relative chemical activity and a measure of their probable resistance to corrosion in the furnace.

Cones sawed from bricks and placed in pats of basic materials were treated in a pot furnace together with a fireclay pat containing standard Orton pyrometric cones. Specimen *A* was found, after a few trials, to react with firebrick between Orton cones 18 and 26. A pat was then heated rapidly to cone 18 and although this temperature was held for 3 hr. no action was observed. Another pat at cone 19 showed the same result. A third pat heated to cone 20 and held for 3 hr. resulted in a violent reaction between the magnesite and firebrick. This procedure was followed in each case.

TABLE 7.—*Reacting Temperatures in Orton Cones between Basic Materials and Brick*

Material	Firebrick	Silica Brick	Lime Content of Basic Material, Per Cent.
A	Cone 20	Cone 26+	3.16
B	Cone 18	Cone 18+	18.96
C	Cone 15	Cone 16	49.19
D	Cone 15	Cone 16	48.15

The figures given in Table 7 illustrate the tendency of lime to lower the reacting temperature and to increase the chemical activity of the basic refractories. It is probable that the temperatures shown are much too high for practical limits of safety, and that they are only relatively true.

The firebrick formed a thin liquid and ran into the magnesite; the shadows in Fig. 2 show where the firebrick had been. The silica brick cones behaved differently and can be seen as they were at the end of the test, having cut away the magnesite without the formation of a visible slag.

Crucible Tests.—Basic refractories, besides withstanding slaking and erosion, must resist considerable chemical action. Carbon, phosphorus, and silicon are present in the metal bath in the form of carbides, phosphides, silicides, or in an oxidized condition. The lining is necessarily exposed, at least locally, to the corrosive influence of silicon and phosphorus in the melt until they are removed. A series of tests was, therefore, conducted for the purpose of studying the action of phosphorus compounds on the six basic materials. Other tests, in which the action of melts containing varying percentages of silicon and phosphorus are studied, are still under way.

The general plan was to make crucibles of the different products, heat them to a certain temperature, introduce a corrosive mixture, allow it to react for equal lengths of time upon the different crucibles, cool, and analyze the melted portion for silica, lime, and magnesia. Since these three oxides could come only from the crucible their presence in the melt would prove chemical action.

Crucible series 1 involved the use of crucibles 3.5 in. (8.8 cm.) in diameter and 2.5 in. (6.2 cm.) in height, made from the materials as received and bonded by means of gum tragacanth. It was impossible to make crucibles of the above type from material *F* so the basic materials were packed in fireclay crucibles 3 in. (7.6 cm.) in diameter and 4 in. (10 cm.) in height in series II, III, and IV. In each case the corrosive mixture added consisted of one part ferrophosphide and two parts ferric oxide, which was used to oxidize the phosphide and was itself reduced

FIG. 2.—PATS SHOWING ACTION OF MAGNESITE UPON FIREBRICK AND SILICA BRICK. I. ORTON CONES IN CLAY PAT. II. SILICA BRICK CONES IN MAGNESITE PAT. III. FIREBRICK CONES IN MAGNESITE PAT.

in the process. The mixture was added to the hot crucibles by means of an iron ladle. At the conclusion of the melting period the crucibles were cooled, sawed vertically through the center, and samples taken from the melt for analysis.

In the first series of tests the crucibles were heated to 1350° C., 100 gm. of the corrosive mixture was added and allowed to react at this temperature 2 hr. The crucibles are shown in Fig. 3. In the second series, the crucibles were heated to 1300° C. when 100 gm. of corrosive mixture was allowed to react upon them for ½ hr.; the action at the end of this time was too slight to be determined. In the third series of tests, the crucibles were heated to 1300° C. and 100 gm. of corrosive mixture was allowed to react at this temperature for ½ hr.; then the temperature of the furnace was increased to 1350° C., 50 more gm. of mixture was added, and the temperature held constant for ½ hr. The extent of the

corrosive action is shown by Fig. 4. The fourth series of tests was conducted by heating the crucibles to 1300° C. and adding 100 gm. of cor-

FIG. 3.—CRUCIBLES USED IN CRUCIBLE SERIES I.

FIG. 4.—CRUCIBLES USED IN CRUCIBLE SERIES III.

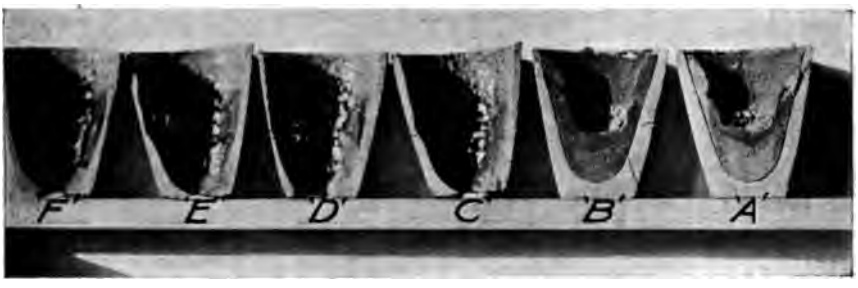

FIG. 5.—CRUCIBLES USED IN CRUCIBLE SERIES IV.

rosive mixture; after ½ hr. the furnace temperature was raised to 1350° and 50 gm. more introduced; ½ hr. later an additional 50 gm. of mix-

ture was added and the temperature maintained ½ hr. longer. Partial analyses of the slags resulting from these tests are given in Table 8.

The analyses of the slags from crucibles *C*, *D*, and *F* in the fourth series should not be compared with those of the same crucibles given in the preceding series, since under the more severe conditions of this test additional reactions came into play. The Fe_3P-Fe_2O_3 mixture had eaten through the bottoms of crucibles *C*, *D*, and *F* as shown in Fig. 5 and a series of reactions resulted between the corrosive mixture, the basic material, and the clay of the crucible, with the formation of a very active fluid slag. The silica, lime, and magnesia contents of this slag were influenced by the action of the clay as well as by that of the phosphorus.

TABLE 8.—*Partial Results of Crucible Tests*

Material	Analyses of Slags from First Series				Analyses of Slags from Third Series				Analyses of Slags from Fourth Series				Order of Resistance to Corrosion
	Silica	Lime	Magnesia	Total	Silica	Lime	Magnesia	Total	Silica	Lime	Magnesia	Total	
A.....	0.00	0.12	1.30	1.42	0.21	0.08	0.35	0.64	0.32	0.04	1.23	1.59	First
B.....	0.31	0.10	1.33	1.74	0.34	0.22	1.55	2.11	0.74	0.46	3.52	4.72	Second
C.....	0.64	0.81	0.69	2.14	2.24	1.42	1.58	5.24	6.20	7.48	4.18	17.86	Fourth
D.....	0.64	1.48	1.47	3.59	9.98	4.55	6.86	21.39	18.90	22.14	14.90	55.94	Sixth
E.....	0.51	0.83	0.64	1.98	1.82	1.13	1.58	4.53	1.56	1.44	1.53	4.53	Third
F.....					1.72	5.37	5.42	12.51	24.52	12.70	8.12	45.34	Fifth

These tests show that the relative corrosion of the various basic materials places the low-lime magnesite first in resistance to the action of the corrosive mixture, the high-lime magnesite second, a calcined dolomite third, a treated dolomite fourth, another calcined dolomite and a treated dolomite fifth and sixth respectively. The materials lowest in lime offered the most resistance to the action of this corrosive mixture. The superiority of the low-lime magnesite, the intermediate position of the high-lime magnesite, and the general lower resistance of the dolomites establish this conclusively. Other oxides as well as lime were taken into the solution but the intensity of this action seems to have been dependent on the amount of lime present. Only limited conclusions can be drawn from these results. Other factors being the same, it should be expected that the basic refractories lowest in lime should give the best service where exposed to the corrosive action of melts in which phosphorus is the principal corrosive agent. However, the relative value of the materials studied can merely be approximated.

Analysis of the slags from the specially treated dolomites gave unexpected results. Material *C* showed slightly greater resistance to corrosion than the calcined impure dolomite *E*; at the same time material *D*, another specially treated product, was decidedly inferior to *E* and *F*, both dolomites that had been calcined without any effort to coat or

protect the granules by special methods. To what extent the special treatment accorded the patented dolomitic products is an advantage in increasing the resistance to corrosion is open to question.

Tests Upon Dead-burned Magnesite and Magnesite Brick

The preceding tests have served to distinguish between magnesites and dolomites. The latter part of the work applies to magnesites only, and the tests were made somewhat more severe in order to secure more pronounced results.

Crucible Tests on Magnesites.—Crucibles made of magnesite grains were subjected to tests similar to those described but under more severe conditions. Crucibles like those employed in the first series were used

Table 9.—*Partial Analyses of Slags From Crucible Tests*

Material	*A*	*B*
Silica	2.56	2.96
Lime	0.40	2.28
Magnesia	2.46	6.38
Total	5.42	11.62

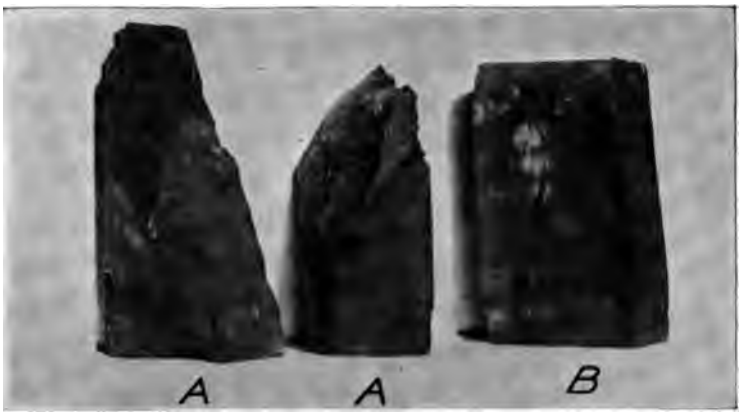

Fig. 6.—*A.* Typical failure of Brick *A* in load test. *B.* Failure of Brick *B* in load test. Original dimensions 8.75 in. by 4.13 in. by 2.53 in.; final dimensions, 7.66 in. by 4.31 in. by 2.72 in.

and 600 gm. of the corrosive mixture added hourly in 150-gm. batches. A temperature of 1350° C. was maintained for 4 hr. The marked superiority of magnesite *A*, in resistance to corrosion, was shown in the preceding experiments and is developed even more strikingly in this one.

Load Tests.—It seems probable that unsatisfactory service would be caused by any marked softening of the bottoms, for the boiling of the bath

will have a greater tendency to scour out holes in a hearth that is soft than in one that is more rigid. In order to study the rigidity of magnesites *A* and *B* at high temperatures, bricks having the compositions given were made from each by the usual brick-making methods, and subjected to a load test. These brick were heated under a load of 25 lb. per sq. in. (1.7 kg. per sq. cm.), applied on the end of the brick with a temperature increase of 250° C. per hr., until failure resulted. Brick *A* failed by shearing at 1555° C., evidently due to a breaking of the bond; brick *B* did not shear but softened and settled at high temperatures. By the time 1450° C. had been reached, it had shortened 12.5 per cent. of its original length, see Fig. 6.

TABLE 10.—*Analyses of Magnesite Bricks Tested*

Material	*A*	*B*
Ignition loss	0.26	0.37
Silica	.7.26	5.72
Alumina	2.14	2.00
Ferric oxide	4.95	5.94
Lime	3.18	15.05
Magnesia	82.40	70.84
Total	100.19	99.92

Since magnesite *A* shows no evidence of softening at the temperature of the test while magnesite *B* shows considerable, it is to be expected that the former should show greater resistance to the erosion of boiling metal than the latter.

Fusion Tests.—The difference in softening temperatures of the two magnesites indicated that fusion-point data might be of value. It was hoped that the fusion points of the magnesites and also of intimate mixtures of finely ground slag and magnesite might be obtained. Were the latter possible, comparative values could be established. One could state, for example, that a mixture of 80 per cent. of magnesite *X* and 20 per cent. of slag was equivalent to a mixture of 60 per cent. of magnesite *Y* and 40 per cent. of slag. Cones were accordingly made up, some of 100 per cent. magnesite and others of magnesite mixed with from 10 to 90 per cent. of basic open-hearth slag.

It was first necessary to devise a method for placing the test pieces in material that would be neutral to both cones and magnesite. After several failures, successful cone pats were made in the following manner: One half of the pat was made of fireclay and Orton cones were placed in this; the other half, in which the test cones were placed, was made of magnesite. The two half pats were joined to form a disk 3 in. in diameter and finely ground chrome ore used at the junction to separate the magne-

site and fireclay. The disk was then placed upon a fireclay pedestal that had been sprinkled with crushed chrome ore. This combination was necessary to prevent the magnesite from attacking the high-temperature Orton cones, or the fireclay of the pat.

When the method for conducting the tests had been devised, it was found that the slag ran from the cones at high temperatures and left a magnesite shell standing. The results of such tests were therefore disappointing. Slightly better but inconclusive results were obtained when ferric oxide was used in place of the slag. Cones of 100 per cent. magnesite also gave poor results, because of the tendency of this material to volatilize at high temperatures.

Bar Tests.—The load test indicated differences in the softening points of magnesites *A* and *B* but the attempted study of fusion points revealed

FIG. 7.—TYPICAL BAR TEST RESULTS.

nothing. It was hoped, therefore, that a series of bar tests, conducted in the following manner, would combine information that might have been secured from the other two.

Magnesite and basic open-hearth slag were ground to pass 60-mesh, weighed out in the proportions indicated, and mixed with gum tragacanth. The mixtures were molded into bars 1 in. by ⅜ in. by 6 in. (2.5 by 0.9 by 15 cm.), which, when dry, were strong enough to be introduced into the furnace. They were laid flat on silica brick and heated to 1250° C., a temperature sufficient to produce a certain amount of sintering. The sintered bars were placed in a test furnace and supported between bricks placed 4½ in. (12 cm.) apart, being separated from the latter by chrome ore powder. The furnace was then heated at the rate of 100° C. temperature increase per hour, and the softening points were observed. The observations are given in Tables 11 and 12.

These mixtures showed no evidence of softening until the temperature of failure had been reached. In the entire series sagging began at a point

TABLE 11.—*Bar-test Data with Magnesite A*

Mixture	Furnace Temperature at which Softening Began, in Degrees C.	Remarks
100% *A*	Samples broke	Too refractory to become bonded by this treatment.
70% *A* + 30% slag	1370	
60% *A* + 40% slag	1350	
50% *A* + 50% slag	1350	
40% *A* + 60% slag	1350	Action complete (sagging of 2 in.) in about 2 min.
30% *A* + 70% slag	1345	
25% *A* + 75% slag	1340	
20% *A* + 80% slag	1330	
15% *A* + 85% slag	1320	
10% *A* + 90% slag	1310	

TABLE 12.—*Bar-test Data with Magnesite B*

Mixture	Furnace Temperature at which Softening and Sagging Began, in Degrees C.	Remarks
100% *B*	1285	These bars sagged about 1 in. in 1 hr. when 1350° C. had been reached.
95% *B* + 5% slag	1280	
90% *B* + 10% slag	1260	Had sagged 2 in. at 1340° C.
85% *B* + 15% slag	1275	
80% *B* + 20% slag	1280	Had sagged 2 in. at 1350° C.
70% *B* + 30% slag	1260	
60% *B* + 40% slag	1250	
50% *B* + 50% slag	1260	

above the fusion point of the slag, which melts at 1290° to 1300° C. All magnesite *B* bars, as shown in Table 12, began to soften about 1 hr. before the action was complete (when they had sagged 2 in.), and at about the melting point of the slag or a trifle lower. The temperatures at which the bars had sagged 2 in. were nearly the same when appreciable amounts of slag were present.

The results of the bar tests are not easy to interpret and it is not possible to state that the softening tendency of a slag-magnesite mixture of one series is equivalent to that of a definite mixture of the other. Failure of the bars evidently occurred through softening and fusion of the slag. The melting temperature of the latter seems to have been raised by admixture with magnesite *A*, and depressed by magnesite *B*. What the significance of the latter fact may be is not clear.

SUMMARY

1. *Comparison of Low-lime and High-lime Magnesite.*—The magnesite that was the lower in lime showed less tendency to slake and higher re-

fractoriness, as well as greater resistance to attack by firebrick and silica brick, and to the action of the corrosive mixture.

2. *Comparison of Dolomitic Materials.*—The materials highest in impurities and lowest in lime were most resistant to slaking. With one specially prepared dolomite *C* in which the granules had been coated with basic open-hearth slag, the inherent tendency of dolomite to slake had been overcome to a great extent. Another special preparation of similar character *D* showed practically the same degree of slaking as the untreated dolomites. The specially treated dolomite *C* withstood the action of the corrosive mixture not quite as well as an untreated dolomite high in impurities, and much better than the second special preparation *D*. Except material *D*, the purest dolomite showed the poorest resistance to corrosion, although this may perhaps be explained by the high ignition loss of the material as received and used.

3. *Comparison of Magnesite and Dolomite.*—The magnesites are more resistant than the dolomites or dolomitic preparations to slaking, also to the action of the corrosive mixture and that of fireclay and silica. One of the specially treated dolomites has a slaking tendency so low as to group it with the magnesites as far as this property is concerned. However, in resistance to corrosion it compares more closely with the untreated calcined dolomite high in impurities.

DISCUSSION

J. S. UNGER,* Pittsburgh, Pa. (written discussion†).—From the subject of the paper it is natural to expect that the data presented must be results secured from an open-hearth furnace working under normal conditions, but such is not the case, as the conclusions are based on the results of laboratory tests, which do not approach service conditions.

The early basic hearths were made of burnt lime, then followed burnt dolomite, bricks made of calcined dolomite and a binder, usually tar, and finally magnesite bricks covered with a layer of sintered magnesite. When this material became scarce, special dolomitic refractories were used.

In the summary, it is claimed that magnesite low in lime, or of greatest purity, ranks first in the resistance to slaking and to corrosive slags. It is well known that neither pure lime nor magnesia has much value as a refractory. By the addition of fluxes, such as silica and iron oxide, lime is rendered inert to slaking in the same way as are the special dolomitic refractories. An example on a large scale is the ordinary basic open-hearth cinder, produced every day, which does not slake. Pure magnesia, regardless of the lime content, is not any better fitted for a refractory than pure lime. The pure natural varieties, such as Grecian or Washing-

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ton magnesites, must be burnt with sufficient silica and oxide of iron to produce a satisfactory material. Grain magnesite, either artificial or natural, is not used in burning in a hearth in the state in which it is received. To make a satisfactory hearth, it must be made more impure by adding about 20 per cent. of basic open-hearth cinder, otherwise it will not remain in place.

Magnesite bricks do not come in contact with the metallic bath or the slag, which is the corrosive agent. Where exposed, they are acted upon by fumes and temperature alone. The magnesite bricks in the hearth are covered with the impure material composing the sintered hearth, which, in spite of a lower softening point, is the better material of the two. The greatest destruction of the basic hearth is at the slag line. Boiling on a bottom cannot be said to be the result of the chemical action of a high phosphoric slag on a basic bottom, as boiling may occur with a very low phosphorus charge not in contact with a basic material, such as a boil in an acid furnace.

It is hard to believe that a slag composed of oxide of iron containing about 6 per cent. of phosphorus is the active corroding agent in a boil in a furnace, especially if a high pig charge is used and the first run-off of slag has removed 90 per cent. of all the phosphorus at a low temperature, in less than 1 hr. after adding the molten metal and sometimes before the charge has completely melted.

When the basic open-hearth steel furnace was first used, it was necessary to work with such charges as had the least action on the hearth lining. The basic open-hearth manager has always insisted that his ore and limestone must be low in silica, which is an active corroding agent. The blast furnaces are operated to furnish a basic iron low in silicon, which oxidizes to silica during the steel-making operation. The scrap carries very little silicon. When the pig-iron percentage is high, a great deal of silicon is carried into the bath and an effort is made to remove it quickly before serious damage is done to the lining. If by accident a high-silicon, low-phosphorus pig is charged on a basic hearth and melted, the hearth in its immediate neighborhood is very quickly destroyed and the results of such corrosion are easily seen after the heat has been tapped.

The experimental slags used to determine the quality of slag resistance of the several basic materials used in the investigation do not represent nor even closely approach the slags found in practice, consequently the conclusions reached may be entirely different from those actually found at the furnace. Magnesite will hardly supersede the ordinary raw or calcined dolomite used in fettling after a heat. Should the cost be brought low enough, pure magnesite would have to be made sufficiently impure to set quickly, otherwise it would not do.

It is very difficult in hearth linings to show any superiority of grain

magnesite over good dolomite refractories. Every factor, such as cost of refractories, labor in patching, continuous operations, and tonnage produced in a given time, must be considered. When everything is taken into account the economies about balance, but the magnesite is preferred principally because it gives less trouble and inconvenience.

FRANCIS R. PYNE,* Chrome, N. J. (written discussion†).—The question of basic refractories, and, in fact, almost all types of refractories, is usually approached from the point of the steel maker alone. However, it should be kept in mind that the non-ferrous metallurgists are vitally interested in the problem of refractories applied to their own work. Refractories satisfactory to the steel man are not necessarily so to the non-ferrous metallurgist, while a refractory satisfactory to the non-ferrous metallurgist would not be so considered by the steel metallurgist. The non-ferrous metallurgists do not have to contend with the high temperature of the steel industry and they have radically different conditions in their furnaces. Notwithstanding these facts, the non-ferrous metallurgist is usually required to select his refractories from those prepared for steel use, often to his financial and metallurgical detriment.

On page 92, the authors state that the use of Canadian magnesite is being abandoned by the leading refractory manufacturers in favor of the California and Washington magnesites, and throughout the paper Canadian magnesite is given a low rating as compared with the low-lime magnesites. While this may be correct for steel work, Canadian magnesite has been found very satisfactory for copper work; in fact, it is nearly on a par with Hungarian magnesite.

RAYMOND M. HOWE (author's reply to discussion‡).—Doctor Unger says that magnesite will hardly supersede dolomite in fettling after a heat; that it is hard to believe a slag composed of oxide of iron containing about 6 per cent. of phosphorus is the active corroding agent in a boil in a furnace; and that boiling cannot be said to be the result of the chemical action of a high-phosphorus slag on a basic bottom. These are well-established facts that no one would venture to question.

The authors were misinterpreted in the statement that "in the summary, it is claimed that magnesite low in lime, or of greatest purity, ranks first in resistance to slaking," etc. On the contrary, the position has been taken throughout that while a low lime content is advantageous, the presence of several per cent. of iron oxide (an impurity) is necessary in order to confer the proper bonding and sintering properties in dead-burned grain magnesite to be used for furnace hearths. We cannot subscribe to the view that pure magnesia is no better fitted for a refractory than pure lime, nor that the pure natural varieties of magnesite,

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† Received Feb. 19, 1919.
‡ Received May 1, 1919.

such as Grecian or Washington, must necessarily be burned with silica and oxide of iron to produce a satisfactory material. On a small scale, pure magnesia has been satisfactorily used as a refractory in the form of crucibles. In the production of dead-burned grain magnesite from the Grecian or Washington mineral, iron oxide is added but the addition of silica is neither customary nor desirable. It is possible to manufacture brick of good quality from both the Grecian and the Washington material without adding either silica or iron oxide, providing they are burned at a sufficiently high temperature. This is regularly done on a commercial scale both in America and abroad when brick of exceptionally high refractoriness or special purity are desired. The following analysis is of an English brick of Grecian magnesite which is extremely low in all impurities.[6]

	Per Cent.
MgO	94.74
Al_2O_3	2.67
Fe_2O_3	Trace
CaO	0.60
SiO_2	1.53
Loss on ignition	0.24
Total	99.78

While 5 to 20 per cent. of basic cinder is added to the dead-burned magnesite in making the furnace hearth, it is probable that there is little if any chemical reaction between the magnesite and the cinder that would render the former more impure. The slag probably acts merely as an agglomerant to hold the grains together.

Dead-burned grain magnesite with a certain percentage of basic cinder added is better suited for use in furnace bottoms than magnesite brick because it can be made into a solid hearth without joints and of any desired contour, which hearth as it wears away can be continually built up by patching with similar material. Although his meaning is not clear, Doctor Unger probably had something like this in mind in describing impure material composing the sintered hearth as a better refractory than magnesite brick, in spite of its lower softening point. From a literal interpretation of the language used, it would logically follow that the quality of magnesite brick would be improved by adding 20 per cent. of basic open-hearth cinder to the dead-burned material from which the brick are manufactured. Such a conclusion is contrary to well-known facts.

That silica is the principal corroding agent in the furnace, and that great damage may be done to the basic lining by a high-silicon charge, are clearly brought out in the discussion. The authors have made

[6] A. B. Searle: "Refractory Materials," 247. London, 1917, Chas. Griffin & Co.

tests to determine the relative resistance of the different basic materials to the action of high-silicon melts, as already indicated on page 100. The study consisted of four series of crucible tests, of which only one will be here described as all four lead to essentially the same conclusions.

This melt consisted of 0.65 per cent. phosphorus, 5.20 per cent. silicon, three-eighths in the form of SiO_2 sand, 2.60 per cent. manganese, sufficient oxygen (from Fe_2O_3) to oxidize the impurities, and the balance metallic iron. The percentages of impurities were intentionally made much higher than they are in an open-hearth charge, in order to intensify the action sufficiently to get measurable results in a short time in the laboratory. Such speeding up of reactions, by the exaggeration of certain conditions, is a proper and necessary method of laboratory procedure.

The slags formed during one hour's exposure, after melting the corrosive mixture described, in crucibles made of the various basic materials, contained the following percentages of CaO plus MgO at the end of the test.

Basic Material

A Per Cent.	*B* Per Cent.	*C* Per Cent.	*D* Per Cent.	*E* Per Cent.
0.80	3.9	4.3	8.3	37.0

In this test, as in all the former ones, the materials low in lime offered the greatest resistance to the action of the corrosive mixture; as in those tests, the low-lime magnesite *A* stands first and the high-lime magnesite *B* second, and both are superior to the dolomites.

The principal advantage of the dolomite preparations for patching the bottom is that they set more quickly than magnesite and patches can therefore be made in less time. The chief disadvantage, aside from the tendency to slake, is that they do not last as long and much more is required. The dolomitic preparations undoubtedly find more general and proper application today as substitutes for dolomite than as substitutes for magnesite. According to the records of one of the leading refractory companies, the consumption of high-grade magnesite in pounds per ton of steel produced is practically back to a normal or pre-war basis in the Pittsburgh district.

Use of Manganese Alloys in Open-hearth Practice*

BY SAMUEL L. HOYT,† E. M., PH., MINNEAPOLIS, MINN.

(New York Meeting, February, 1919)

THE present report represents that part of the work that has been done by the War Minerals Investigation, Manganese Section, of the Bureau of Mines, on the use of manganese alloys in open-hearth practice. The magnitude of the work and the number of people participating in one way or another make it difficult to acknowledge adequately the services of all who have contributed to the success of the work. It would be amiss, however, to pass to the body of the report without mentioning the hearty coöperation and willingness to assist of the various manufacturing interests. With such a sympathetic attitude toward this investigation, our work was assured of success. Inasmuch as the unhappy circumstances that rendered this investigation necessary have now passed, this report is made in the hope that the results may prove of some permanent value to the peace-time operation of our open-hearth industry.

Broadly speaking, the purpose of the manganese investigation was to consider the most suitable means of utilizing our domestic supply of manganese and thus relieve, to the maximum extent possible, our shipping of the burden of importing manganese ores. It was not held that any decided shortage of manganese was imminent but rather that every legitimate means should be utilized for making our ship tonnage available to European service. While we have considerable metallic manganese in this country, it occurs largely as a low-grade mixture with iron or silica or with both. The open-hearth investigations, therefore, were carried on to determine the extent to which domestic or low-grade alloys could properly be substituted for high-grade alloys without materially impairing the steel production as to either quality or quantity. Also, it was held that such an investigation would yield valuable results to the steel industry as well as contribute, in no small way, toward directing future investigations in the same field. To the practical steel manufacturer, especially to him who felt no immediate concern about

* Published by permission of the Director of the U. S. Bureau of Mines. Report of research under the joint auspices of the U. S. Bureau of Mines and the National Research Council.

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a manganese shortage, these last two points appeared to have a special significance. Another and quite obvious portion of the problem lay in the determination of the relative manganese conservation of the different practices, although any decided relief would more likely come from the other direction.

It was recognized at the start that considerable experience along our lines of attack had been gained at individual plants and that a compilation and digest of such results, supplemented possibly by our own investigations, would be the quickest method of approach. This, to a certain extent, controlled the selection of the actual steel plants at which it was desired to carry out the detailed investigations.

The evolution of the details of a definite experimental program from the statement of the general problem, considering the time element and the many and varied factors involved, was not lightly undertaken nor worked out without due consideration of competent metallurgical advice. After a preliminary survey, it seemed important to determine the conditions in open-hearth practice that would lead to a conservation of manganese, both during the working of the heat and in making the final additions; the most satisfactory metallurgical conditions for the use of manganese in the form of low-grade or special alloys; and the effect on the finished steel, as to both quality and "condition," of the various methods and processes studied. With these points in mind, steel plants were selected in which investigations bearing on one or more of these points could be carried out. To accomplish this, it was decided to determine slag and metal compositions during the refining of the heat and to observe the temperature each time a sample was taken, the latter to determine, if possible, the temperature effect. The recovery of manganese was to be determined from the residual and final manganese contents and the weight of the metal. To this end we took a sample of the finished steel, during teeming. By taking samples, at the beginning, toward the middle, and at the end of teeming, we were able to test for non-uniformity. This practice was generally observed throughout the investigation. The data secured were also supplemented by the plant records covering the heats in question as well as by our own observations made during refining, pouring, and teeming.

When planning the steps that should be taken to determine the quality and condition of the steel, it was found that no definite and well-proved method that could be adopted was available. True, the open-hearth melter knows whether his heat is in proper condition but we needed a quantitative estimate of "condition." Without attempting to enter upon a discussion of the physical chemistry of a heat of molten steel, we may say that the condition of a heat of steel must depend, aside from the temperature, on the composition of the steel in those substances that affect the condition. Of these there are two

kinds: (1) those that promote "openness," or the gases, which may be classified as gases that are products of bath reactions, which so far as we know are CO and possibly CO_2, and gases that are absorbed from the furnace gases, such as H, N, CO, and CO_2; and (2) substances that promote "soundness," such as the reducing and solidifying agents, C, Mn, Si, Al, etc. In general, it is held that Mn, Si, and Al inhibit the chemical reactions producing CO by reducing (or partly reducing) the principal constituent that produces the reaction, FeO.[1] The reduction of FeO is the principal means of "settling up" the liquid steel and it is for this reason that Mn is added in the finals. It is also held that Si and Al produce solidity in the finished steel, aside from reducing FeO and CO, either by keeping the hydrogen, nitrogen, etc., in solid solution or by preventing the dissociation of the compounds of those elements and iron.

The obvious procedure to get a quantitative estimate of the condition of the steel, considering both the behavior of the molten metal and the character of the ingot, would be to determine the amounts of the constituents in each of these two groups and to weigh one set against the other. Even this procedure would not, at present, lead to results which could be interpreted with entire confidence, even though there were no uncertainties in the analytic methods, inasmuch as we do not know the quantitative effect of each constituent either by itself or when associated with other constituents in varying amounts. In view of this uncertainty (lack of fundamental data), it was decided to make the analyses and use the results in a qualitative way, at least, to compare the different practices investigated.

Functions of Manganese

During such a critical period as that which has now passed, the question might be raised as to the possibility of eliminating manganese from steel making. This point was considered but it was at once held that the use of manganese is not merely an expedient, one for which some substitute might readily be made, but one of the basic requirements of successful, present-day, steel works' practice. It is quite true that in many cases the actual amount of manganese used in a heat of steel is greater than the purely metallurgical considerations would call for and any excess that might be used may well be considered as so much wasted.

The first function of manganese, broadly considered, is to refine and

[1] In this statement only the metal bath is considered and it is held that FeO, and not Fe_2O_3, is in solution in the steel. According to this idea, reducing action on a slag containing Fe_2O_3 will produce FeO, part of which will enter the steel to react later with C, Mn, and other reducing agents present.

"settle up" the molten bath of steel. The aim is to put the metal in a proper condition for pouring and to produce ingots (or castings) of the desired quality and texture. While manganese is not the most efficient element that can be used for this purpose, calculated from the heat of combustion of the element to its oxide, it is the most satisfactory on account of the excellent condition (freedom from objectionable foreign inclusions) in which it leaves the bath. The theoretical amount of manganese required might possibly be calculated from the amount of oxygen converted from the active form FeO to the inactive form MnO. Assuming an oxygen content of 0.075 per cent. in the unsettled steel and of 0.015 per cent. in the finished steel (oxygen by the Ledebur method), the amount of manganese used in this way will be 0.2 per cent.[2] The amount of manganese required naturally will vary with the condition of the bath and, in order to insure efficient "deoxidation," will be somewhat in excess of the calculated amount. A well-made heat of steel will probably not require more than 0.35 per cent. of manganese.

Manganese is also desirable in steel to improve the rolling properties, in which capacity it appears to serve a dual purpose. First, the manganese deoxidizes and refines the molten steel in such a way as to give ingots of the desired texture without robbing the steel of its hot-working properties. Such ingots may be rolled into finished shape without the formation, in excessive amount, of fissures or of surface defects. Other reducing agents, such as aluminum, silicon, etc., are prone to leave the metal in a poor condition for rolling and forging; while they eliminate one cause of hot shortness, iron oxide, they fail to convert the sulfur into a harmless form, as does manganese, and leave behind their highly refractory oxides, both of which tend to produce poor rolling qualities. Second, manganese, by retarding the rate of coalescence, or grain growth, renders steel less sensitive to the effects of the high rolling temperatures and is supposed to promote plasticity, at least in ordinary steels, at rolling temperatures. Silicon and aluminum, on the other hand, increase, rather than decrease, the grain size of steel. The amount of manganese required in this capacity probably does not exceed 0.35 per cent. in well-made steel. Finally, manganese is desired in the finished steel to secure certain physical or mechanical properties or to make the steel more amenable to subsequent heat treatment.

[2] The writer is informed by the Bureau of Standards that such a calculation is premature, due to our lack of knowledge on the subject of deoxidation and the faultiness of the Ledebur determinations. However, it would seem to the writer, from the work done at the Bureau of Standards ("A Critical Study of the Ledebur Method for Determining Oxygen in Iron and Steel" by J. R. Cain and Earl Pettijohn) that the amount of oxygen determined is the amount present as FeO (active form) subject possibly to an error due to partial reduction of CO during the determination. At any rate, the above is advanced as at least the first approximation of the amount of manganese required simply for destroying the ferrous oxide present in the bath.

Manganese is a very valuable element to the steel industry and its use cannot lightly be dispensed with. It is true that material, such as American Ingot Iron, can be successfully rolled even though no manganese is added, but greater time and care are required. Also, even though higher carbon steels with low manganese can be as successfully rolled, it would be impossible, in so far as we now know, to secure a satisfactory substitute for manganese, which is abundant in this country, to perform all three functions.[3]

Manganese conservation, viewed from this angle, will best be secured by eliminating the manganese specification except in case the amount of manganese present in the finished steel has some definite bearing on the properties or heat treatment of the steel. In other words, whenever only casting and rolling-mill practice (plant problems) are involved, the steel man should be allowed to exercise his own judgment as to the amount of manganese that should be used to give the most satisfactory and economical practice and the finish and quality of the product should be controlled by adequate inspection. On the basis of Mr. Ellicott's figures in *Iron Age* for June 6, 1918, by reducing the manganese requirements by 0.2 per cent. in making plates and shapes and other low-carbon steel (estimated tonnage 21,350,000), 54,000 tons of 80 per cent. ferromanganese could be saved.

Recommendations for Utilization of Domestic Alloys

As a result of this investigation, it is advanced that there are three practices for utilizing our domestic alloys in open-hearth practice that commend themselves; they are as follows, but not in the order of their importance:

1. The use of a molten spiegel mixture for deoxidation and recarburization.

2. The practice of melting and refining the steel bath so as to secure a comparatively high residual manganese content, about 0.3 per cent. manganese.

3. The use of manganese alloys containing silicon.

In selecting plants for investigating these practices, two points were kept in mind. The plant should have either "ordinary" practice, for the sake of comparison, or else one of the three just mentioned and the product or kind of steel made should be representative of the larger tonnages, such as shell steel, plates, sections, etc.

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[3] It is interesting to note that manganese, coming in the periodic system between iron and the strengthening elements on one side and the hardening elements on the other, serves the dual capacity of strengthening and hardening steel, which is not done by any other element.

Molten Spiegel Mixture Practice

The practice has been adopted, at a few plants, of combining in one operation both recarburization and deoxidation by using a mixture of pig iron and spiegel, which has been premelted in a cupola. This molten spiegel mixture contains from 5 to 11 per cent. manganese, 4 per cent. carbon, and the desired amount of silicon, and is added to the ladle during the tapping time in such a way as to get a thorough and uniform mixture of the two streams.

The principal advantages, not considering questions of plant and operating economy, are: A low-grade or domestic alloy can be used in the preparation of the mixture. The deoxidation is accomplished by means of a dilute solution with a consequent increase (on theoretical grounds) of the efficiency of the deoxidizer; this point will receive further consideration later. The deoxidizer is added in the molten state, securing thereby the attendant advantages of this practice, which will be considered at greater length. The amount of the recarburizer is comparatively large and the capacity of the plant is materially (and economically) increased thereby;[4] this is an advantage if a large steel output is desired. Compared to the usual practice of adding carbon and manganese, there should be less likelihood of missing a heat.

This practice, at least at the plant visited and it is understood to be the same elsewhere, is limited to the manufacture of the high-carbon steels, or those running 0.30 per cent. carbon or above. To make steels with 0.20 per cent. carbon would require the working of the carbon to about 0.10 per cent. carbon and the molten addition would have to contain about 20 per cent. manganese (spiegel). The amount of the addition would be reduced from 13,000 lb. (5896 kg.) to about 4000 lb. (1814 kg.) which would mean that some of the advantages just enumerated would decrease in weight and with the increased loss of manganese would probably mean that the practice would no longer be commercially feasible. However, when the other alternative, that is, the use of ferromanganese, either solid or liquid, is considered, the practice of premelting spiegel in the cupola would seem to commend itself as worthy of consideration. In case of undue shortage of high-grade ferromanganese, there can be no doubt that the practice would offer a ready solution of the problem of using domestic alloys in making steel for shapes, plates, etc. Against the increased cost of production, compared to cold ferromanganese practice, there would be the greater uniformity of the product and the more uniform practice.

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[4] There is some doubt as to the propriety of including the fourth as an advantage of this particular practice. The use of pig iron as a recarburizer may be accomplished in other ways with the same economy and increase in plant capacity.

High Residual Manganese Practice

At certain plants the practice of preferential oxidation and elimination of carbon and phosphorus has been developed by means of which the residual manganese is kept at a comparatively high value—about 0.25 to 0.30 per cent., as compared to 0.10 per cent. manganese, for a final carbon content of 0.10 per cent. This is accomplished, broadly speaking, by rapidly removing the phosphorus and retaining it as stable calcium phosphate during the earlier and colder period of melting, by maintaining a high finishing temperature and working the charge with a high manganese content so that the slag contains about 8 per cent. manganese, and by increasing the lime content of the slag to about 47 per cent. as a minimum.

This practice possesses undoubted advantages but they are probably best appreciated by those who have developed the practice and have it in operation on a sound commercial basis. First of all, correctly applied, it leads to the production of high-grade and uniform steel, which means increased rolling-mill practice, fewer rejections, and a more ready market. This is largely due to the fact that the steel is made, where it should be made, in the furnace. A second advantage is derived from the high MnO and CaO contents of the slag; manganese finals can be added to the furnace with a recovery that compares favorably with that of ladle additions. A third advantage is that the same pig iron used for the charge, which contains appreciably more manganese than ordinary basic iron, can be used to recarburize and partly deoxidize the bath; the remainder of the manganese is added as ferromanganese. At a steel plant that operates in conjunction with a blast-furnace plant, a harmonious and economical cycle of plant operations is made possible. At the same time the open-hearth slag can be resmelted in the blast furnace for the recovery of the iron and manganese and the utilization of the lime.

This practice is largely dependent on the amount of phosphorus in the slag, for it would not be worth while to recover the manganese at the expense of unduly increasing the phosphorus content of the pig iron. In this country we are fortunate in having a large amount of quite low-phosphorus ore available. No definite figure can be given as to the maximum allowable phosphorus content of the pig iron, but it is the opinion of at least one steel man who followed this practice that 0.6 per cent. phosphorus would not be prohibitive.

Under the conditions that prevailed during the past year, this practice possessed the additional advantages that high-manganese pig iron could be secured by smelting domestic manganiferous iron ore and that the manganese alloy added to the furnace at the end of the heat could as well be spiegel as ferromanganese, assuming that the finished steel

contains above about 0.10 per cent. carbon. This would not be without its disadvantages, one in particular being that the carbon content of the bath would have to be worked to a lower figure than in present practice. On account of the high cost of spiegel and the greater time required, it is doubtful whether the steel plants will substitute spiegel for ferromanganese. Another point in connection with the possibility of utilizing domestic manganiferous iron ore is that low silica ore can be added to the slag as a source of manganese oxide.

The high manganese content of the charge is generally secured by using a high-manganese pig iron, that is 2 or 3 per cent. of manganese; but it may also be secured by adding manganese ore to the slag or manganese alloys to the bath or by a combination of these methods. This point would be determined by plant economy but it seems doubtful whether the practice would be worth while unless a high-manganese pig iron were available.[5] The loss of this manganese, by which is meant its oxidation and transference to the slag, is quite great. This loss may be kept at a minimum by increasing the basicity of the slag in CaO and FeO, which, combined with the MnO which also acts as a base, exert the desired effect on the manganese of the bath. As the working of the charge progresses, its temperature rises until finally, with the high CaO, and particularly MnO, content of the slag, the carbon is eliminated more rapidly than the manganese, with the result that the manganese can be held to about 0.3 per cent. at the end of the heat. Present data indicate, unfortunately, that no material decrease in the amount of manganese required and no material increase in the recovery of manganese in the additions may be expected, so that the advantages are derived not from a decreased consumption but from the form in which it can be added. Data for one such heat showed that a total of 3728 lb. (1690 kg.) of manganese were used in one form or other to produce 1217 lb. (552 kg.) of manganese in the finished steel; or in other words that 3.06 lb. (1.38 kg.) of manganese were used to produce 1 lb. of manganese in the finished steel. The manganese added in the recarburizer and ferromanganese amounted to 1068 lb. (484 kg.) of which, assuming the manganese loss to come from these two sources, 838 lb. (380 kg.) were recovered in the finished steel, a recovery of 78.4 per cent. In this heat the ferromanganese was added to the furnace. Another heat selected at random, but more representative of standard practice, used 2190 lb. (993 kg.) of manganese to produce 1200 lb. (544 kg.) of manganese in the finished steel, or 1.82 lb. (0.8 kg.) of manganese (as compared to 3.06 lb.) to produce 1 lb. of manganese in the finished steel.

[5] The writer is informed by one blast-furnace superintendent that running the manganese up to 2 per cent. does not materially affect the production, so that lowering of pig-iron production would not be held as a disadvantage in this practice.

Open-hearth Investigations

Open-hearth investigations, in which standard practice will be compared to molten spiegel practice and high residual manganese practice, will be described later after more complete analytic results have been received.

Use of Manganese-silicon Alloys

The high silica content of most of our domestic manganese and manganiferous iron ores made it advisable to consider the possible use of manganese-silicon alloys in steel making, in both acid and basic practice. For the purposes of the present discussion, these alloys will be divided roughly into two classes, high-grade silico-manganese containing about 50 per cent. manganese and 25 per cent. silicon, and low-grade silico-spiegel with about 15 to 20 per cent. silicon and 30 to 35 per cent. manganese with 50 per cent. iron. The manganese-silicon ratio of the first alloy is about 2 and of the second alloy varies from 2⅓ to 1½. Both of these alloys contain very little carbon. The high-grade alloys would be made from the siliceous manganese ores of California and Montana, and the low-grade alloys from the siliceous manganiferous iron ores of Minnesota.

While there is nothing new about the practice of using manganese-silicon alloys in steel making,[6] it may be well to review some of the points connected therewith in the light of present requirements. Silicon is always an efficient reducing or settling up agent when in the customary small amounts, but it may or may not be desirable in the finished steel. On this account the possibility of using manganese-silicon alloys depends on the amount of silicon that can be tolerated in the finished steel in the ingot form. In certain grades of steel, particularly in steel that must be welded, silicon should be low or practically absent. In steel for sheets and plates, which must give a good finished surface, most efficient rolling-mill practice requires that the silicon be kept tolerably low, but it is believed that from 0.10 to 0.15 per cent. could be used provided the manganese were not too high. In forging steel, high-carbon steels, and castings, where the aim is to produce sound steel, more silicon can be used, or between 0.20 and 0.35 per cent. Of these three fields, the last is the one where manganese-silicon alloys will find their first application. In the second field, it seems quite probable that conditions (to be discussed later) will many times permit their use; but from the very nature of things manganese-silicon alloys cannot be used to make steels of the first group.

[6] It is understood that "silico-manganese" has been used fairly extensively in Europe, and in this country it was used at certain plants as standard practice until the supply was cut off by the war.

Manganese-silicon Alloys in Acid Practice

It is with considerable diffidence and hesitation that the discussion of manganese-silicon alloys in open-hearth practice is approached, particularly as the controversial character of many of the points is so clearly recognized. So it may be well, at the outset, to state briefly the manner in which the writer became interested in the possibilities of their use. A number of years ago, he was conducting a series of experiments on the occurrence and identification of foreign inclusions in acid open-hearth steel, principally ordnance steel. In this work ferromanganese, ferrosilicon, and a mixture of ferromanganese and ferrosilicon were added to a steel sample taken shortly after "oreing," that is to "wild" steel, in an attempt to produce an excess of the constituent, or constituents, that were supposed to form as a result of the addition. It seemed fairly clear as a result of this work that the use of silicon was apt to be dangerous, not on account of any harmful effect of the residual metallic silicon but because it produced a constituent (assumed to be SiO_2 or at least a highly refractory silicate) that was very likely to remain in the ingot and produce hot shortness. This suggested the idea that a manganese-silicon alloy might, and probably would, form a manganese silicate containing some ferrous oxide (a true slag) that would be fluid and more readily coalesce into larger particles than SiO_2, and, therefore, would free itself more readily from the steel. By using such an alloy, it would be possible to take full advantage of the use of silicon as a deoxidizer without suffering its usual attendant disadvantages. None of the manganese-silicon alloy was available at the time, so a parallel experiment could not be conducted.

Aside from the possibility of securing a better separation of the insoluble products of the deoxidation process, it was assumed as likely, from the fact that binary alloys are known to be generally more active, or powerful, than the weighted sum of the two constituents would indicate, that the alloy of manganese and silicon would prove to be a more powerful reducing agent than ferromanganese and ferrosilicon used separately. On reflection the thought occurs that manganese and silicon, reacting separately with FeO, would produce the oxides MnO or SiO_2, which may or may not form a solution of MnO and FeO or a silicate of iron. Manganese and silicon reacting as an alloy with FeO would produce a silicate of manganese, which may or may not form a double silicate with FeO. In either case we would expect to find the advantage in favor of the manganese-silicon alloy. The relative weight of the silico-manganese and the ferromanganese plus ferrosilicon mixture will be considered elsewhere.

Another point of great technical importance is the percentage recovery of manganese when added as silico-manganese and as ferromanganese

along with ferrosilicon. First of all let it be stated that a 100 per cent. recovery, based on the present theory of deoxidation, is hardly possible; and if possible would not be desirable. It would mean a retention of the products of the deoxidation, to be determined later as metallic manganese and silicon. A method of addition that would lead to a satisfactory deoxidation and yet would eliminate the loss due to admixture with the slag, volatilization, etc. and could be accomplished with the minimum amount of manganese, would be very desirable because it would lead to both conservation of manganese and uniformity of composition of the steel. The first of these points would be given by the actual value of the percentage recovery of the manganese and the second by the constancy of the percentage recovery.

Fortunately, the writer was able to examine records of heats made with silico-manganese covering a period of several years, from which some fairly satisfactory conclusions may be drawn bearing on these points. During this time when silico-manganese was being used there were periods when the alloy was not available and the ferromanganese plus ferrosilicon mixture had to be substituted. This afforded a direct comparison of these two methods of deoxidation. Certain results taken from the heat records, and which are believed to be typical,[7] are given in Table 1. Approximately, the silico-manganese contained

TABLE 1.—*Silico-manganese vs. Ferromanganese + Ferrosilicon*

Heat	FeSi, Pounds	FeMn, Pounds	SiMn, Pounds	Total Charge, Pounds	C, Per Cent.	Mn, Per Cent.	Si, Per Cent.	Mn Added, Pounds	Mn Recovered, Pounds	Per Cent. Recovery
A	160	300	...	31,160	0.26	0.56	0.294	240	174	72.5
B	215	400	...	40,765	0.21	0.57	0.306	320	232	72.5
C	160	310	...	30,620	0.32	0.63	0.312	248	193	77.8
D	160	300	...	30,910	0.27	0.70	0.318	240	216	90.0
E	215	400	...	40,865	0.24	0.72	0.312	320	294	91.8
F	...	40	470	41,010	0.22	0.58	0.308	281	238	84.7
G	...	35	350	30,685	0.26	0.60	0.302	214	184	86.0
H	...	40	470	40,960	0.21	0.64	0.310	281	261	93.0
I	...	40	470	40,510	0.21	0.66	0.308	281	267	95.0
J	...	40	420	36,760	0.24	0.68	0.310	255	250	98.0

53 per cent. manganese and 20 per cent. silicon; the ferromanganese, 80 per cent. manganese; and the ferrosilicon, 50 per cent. silicon. The residual manganese was neglected in calculating recoveries.

[7] It is obvious that the variation in heat composition and average manganese recovery of several years' practice cannot be given by this table. The records show greater uniformity for the silico-manganese heats.

It can hardly be claimed for these figures, or for the 3 years' records they represent with reasonable accuracy, that they furnish a truly scientific basis of comparison of the two alternate practices, but they do show that the same results (manganese and silicon contents of the finished steel), by using silico-manganese, can be obtained with consistently smaller amounts of both manganese and silicon, as compared to the combination of ferromanganese and ferrosilicon. In addition, there is the advantage of having a more uniform practice, which in itself would warrant smaller additions. The weights of the additions favor the silico-manganese; thus in heats A and D, 460 lb. (208 kg.) and in heat C 470 lb. (213 kg.) were added as compared to 385 lb. (174 kg.) for G; and in B and E, 615 lb. (278 kg.) were added as compared to 510 lb. (231 kg.) in F, H, and I. The low carbon content of the silico-manganese may or may not be a material advantage but it is in favor of the single alloy addition because the carbon need not be worked as low and there seems to be less danger of missing the carbon.

Manganese-silicon Alloys in Electric-furnace Practice

No information is available, to the writer, bearing on the use of these alloys in electric-furnace practice, but we may at least consider such a possibility on the basis of their known behavior. Considering acid casting practice first, there seems to be no reasonable doubt that either silico-manganese or silico-spiegel could be at once substituted for ferromanganese and ferrosilicon. Inasmuch as the usual aim is to make high-grade castings, the manganese-silicon alloys would appear to have the distinct advantage of making sounder and cleaner steel. Silico-spiegel, aside from possessing the theoretical advantage of being diluted with iron,[8] could be more readily prepared with the correct manganese-silicon ratio so as to eliminate the use of an additional alloy. The uncertainty of our knowledge as to the relative behavior of the manganese-silicon alloys as compared to the ferro-alloys, and the relative efficiency of low-grade and high-grade alloys, as well as the importance of this step in the manufacture of steel, suggest the advisability of conducting a definite research to settle the points. It would seem that there is no better place for such a research than in this particular industry.

In basic electric-furnace practice the manganese-silicon alloys, on the same grounds, could likewise be utilized, particularly as the attempt is always to produce sound and clean ingots. However, in this practice, ferrosilicon is used as a reducing agent along with coke and hence the

[8] In this practice the advantage of greater dilution need not carry with it the disadvantage of increased weight on account of the higher temperature of the electric furnace.

operator would probably not see any advantage in changing his practice in favor of the manganese-silicon alloys.

Manganese-silicon Alloys in Basic Open-hearth Practice

The amount of information available on the use of silico-manganese in basic open-hearth practice is very meager, but it can be said that silico-manganese would probably be as satisfactory as ferromanganese and ferrosilicon. Through the coöperation of one steel plant, we were able to follow two shell-steel heats made with silico-manganese, which was added to the ladle. The second of these heats is given here to show what was done. To 11,100 lb. (5034 kg.) of molten pig iron in the ladle were tapped 122,340 lb. (estimated) (55,492 kg.) of steel analyzing C, 0.09, P, 0.012, Mn, 0.15, S, 0.033, Si, 0.02. At the same time 1000 lb. (453 kg.) of silico-manganese (50 per cent. manganese), 300 lb. (136 kg.) of 70 per cent. ferromanganese, 12 lb. (5 kg.) of aluminum, and 50 lb. (22 kg.) of coal were added to the ladle. The heat was in excellent condition and the ingots had smooth even tops and displayed no superficial action or evolution of gas. The final analysis was C, 0.44, P, 0.028, Mn, 0.58, S, 0.041, and Si, 0.21. The recovery of manganese, assuming the entire loss to come from the alloy added, was 77.5 per cent.; the recovery of silicon was 65.1 per cent. Only 24 lb. (10 kg.) of carbon, or 5 per cent. of the total, was lost. In the first heat, which was thought to be more highly oxidized, the recovery of silicon was only 58 per cent., while the recovery of manganese was 73.4 per cent. This indicates that silicon protects manganese in oxidized heats.

It was our intention to follow two more heats made by catching the carbon coming down and by adding the silico-manganese to the furnace just before tapping, but this has not as yet been done.

DISCUSSION

Samuel L. Hoyt.—The question of adding the ferromanganese to the ladle or to the furnace involves both theoretical and practical questions and its discussion might very easily occupy the rest of the day. One of the points to be kept in mind is the desirability of obtaining uniform steel. If ferromanganese is added to the ladle, it is necessary first to melt the manganese and then to distribute uniformly throughout the heat. The only heat available for this purpose is that of the molten steel itself. No doubt sufficient heat is present, but in our desire to pour the steel as quickly as possible there is hardly sufficient time allowed, many times, for the proper distribution of the manganese. In other words, this is an operation which takes a certain amount of time and any undue shortening of the time allowed must affect the quality and uniformity of the steel.

Again, this particular method does not lead to as efficient a use of the manganese as might be desired; a contention which is based on theoretical grounds. In concentrated solution, manganese will completely reduce all the ferrous oxide, the presence of which I hypothecate and produce, as the reaction product, manganese oxide. In dilute solution, the manganese only partly reduces the iron oxide, but the manganese oxide which does form dissolves the iron oxide in simple solution. The dissolution of ferrous oxide in the manganese oxide will result in a more economical utilization of the manganese and an equal purification of the heat. This whole point, at present, has not been definitely proved and its answer will involve very difficult research work.

On the other hand, if the manganese is added to the furnace, the heat of the furnace is used to melt the alloy and there is plenty of time during tapping for the manganese to be very uniformly distributed throughout the steel. By distributing the alloy throughout the charge we can secure a better utilization of the manganese than is possible if the addition is made to the ladle. Certain other points, such as the recovery, have been dealt with at greater length in the paper.

The practice of adding manganese in the molten condition to the ladle commends itself for two reasons, considered in the light of the present discussion. It is possible to pour the molten ferromanganese or molten spiegel mixture in such a manner that it unites with the steel stream so that at no time is there a considerable excess of manganese. This insures a proper utilization of the manganese. Furthermore, the manganese, at the same time, is distributed uniformly throughout the heat.

THE CHAIRMAN (HENRY M. HOWE, Bedford Hills, N. Y.).—We must admit that at first sight the addition of ferromanganese in the ladle seems to be an economy; less of it is required and there is less opportunity for oxidation and loss. But if you are making cannon or other superior steel and want high quality there can be no doubt that the addition should be made in the furnace. One very important point is that the steel in the ladle should not be completely molten. This is shown by the fact that all good practice requires that steel of any good quality should leave a considerable skull in the ladle, and, of course, the metal which forms this skull is not only below the liquidus but below the solidus. This almost necessarily implies, in view of the pretty wide gap between the solidus and the liquidus, that the steel as a whole is below its liquidus. This in turn implies that it is not a true liquid, but an emulsion of particles of solid steel suspended in a mass of molten steel.

If under these conditions you throw cold solid metal into the ladle it is probable, but not very easy to show, that the emulsion will immediately solidify around the cold ferromanganese and form a lump, perhaps the size of your head, of ferromanganese on the inside and solid steel on

the outside. Then before it is possible to diffuse and distribute that manganese it is necessary to melt off the solid steel to get at the manganese. That I think is the great consideration. If you are going in simply for economy and to make steel as cheaply as possible, and if the cheapest steel is as good as the public needs, by all means add the ferromanganese in the ladle; but if you want to make really good steel add the ferromanganese in the furnace.

FRANK N. SPELLER,* Pittsburgh, Pa.—The addition of manganese to the open-hearth bath does not always give the most beneficial results in the finished product, depending on what operations the steel is put through in fabrication. To get the cleanest steel, no doubt additions to the furnace give the best results due to the more thorough mixture and more complete removal of byproducts of deoxidation; but on the other hand this tends to produce a "dry" steel, which gives rise to a more rapid crystalline growth and tendency to overheated structure in subsequent heating. This is what we find in making welds with steel too highly finished either by this or other means. It is unsafe, therefore, to accept the statement of previous speakers without reservations; viz., that to get the best results the manganese additions should not be made in the ladle.

HENRY TRAPHAGEN,† Toledo, Ohio.—Professor Hoyt has demonstrated that a satisfactory manganese content can be obtained in open-hearth steel by the use of these low-grade manganese silicon alloys. In fact, the claim seems to be established that the manganese return from this material is higher than usual. Does this mean that the higher manganese return is due to better than usual melting conditions, or does it mean that the manganese in these alloys is below par in deoxidizing power? I rather incline toward the latter view, for I have invariably found, in practice, that the lower grades of manganese alloys were deficient in this respect. They usually show a high yield but fail to properly kill the bath.

The heats under Professor Hoyt's management may have been made under ideal conditions and the amount of oxidation kept at a minimum. If this were so, then 80 per cent. ferromanganese should have shown a greater than usual yield, if it had been used. But if these low-grade alloys consistently show a higher return than the standard ferromanganese, then look to your steel. Be certain that it is up to par, not in one instance, but throughout a long series of test heats, before accepting low-grade ferromanganese as the solution of manganese conservation.

There is no doubt that this manganese question will be solved. But when it is solved, we shall certainly change our ideas regarding

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the amount of manganese necessary in good steel, for we will find that the solution lies in good raw materials carefully melted, and that true conservation will result from the knowledge that very much less ferromanganese per heat is required. But this will never happen until quality becomes our national watchword, and the present-day "pay by the ton, get the stuff out" policy is no more.

S. L. HOYT.—I would like to ask for a little further consideration of that last point. Would you mind stating the reason why the higher percentage of manganese should give better steel?

HENRY TRAPHAGEN.—I base my remarks on the statement that the recovery of manganese from this low-grade material is so high. It appears to me that if the recovery is very high the manganese has not been doing its work. Personally, I would rather have the manganese lost under certain conditions than have it go into the steel. By that I do not mean that I would rather see the manganese thrown into the slag and carried out into the dump, but I would like to see the manganese utilized in cleansing and quieting the steel, rather than see it passed into the metal along with a mass of undecomposed oxides.

C. L. KINNEY, JR.,* South Chicago, Ill. (written discussion†).—It is obvious that the subject matter of Mr. Hoyt's comprehensive paper can only be briefly discussed upon one basis, namely, quality of product; and this basis is selected in the conviction that the steel industry will be confronted, year by year, with an ever-increasing need of meeting more difficult physical specifications. It follows, then, that no procedure should be considered which in any manner handicaps the producer in meeting these inevitable demands. Conversely, any method that will, in a simple economical manner, produce improvements in quality should be given careful consideration. Further, no method or process to which is attached even a suspicion of sacrifice of quality in the name of production, utilization, or conservation should be considered in any light other than that of commercial retrogression.

I believe it is generally recognized among open-hearth operators that there is only one place to attain quality, and that is in the furnace. True, the steel can be affected by improper pit operation and by careless heating and rolling; but in the main it seems the ever-increasing duty of the steel producer to send to the finishing mills a product that can successfully withstand the stresses imposed upon it by modern mill practice.

It now becomes a question of method of attaining this degree of quality by practices applicable to plants producing large tonnages of

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†Received, Feb. 15, 1919.

basic open-hearth steel. Preëminent among all expedients stands the necessity for thorough deoxidation, and the correlated fact that its products must be given opportunity to escape from the steel. Can one logically assume that these demands are met when a bath of steel, low in residual manganese and covered by a basic slag carrying the general percentage of oxides, is deoxidized by the almost archaic method of hurriedly adding a few hundred pounds of deoxidizer in the ladle and inevitably pouring the steel almost immediately? I am sure that the answer is "No." It therefore follows that deoxidation should not be the grand finale of the process. Rather, it must be an integral part, functioning continuously from the period when the bath is melted and slag formation completed, up to the time of tapping, and at this time such further additions may be made as are indicated by the chemical composition sought. Under these conditions there is apparently but one metal that will function properly and economically, and that is manganese.

Assuming the preceding hypotheses to be sound, the problem becomes: "in what form can this manganese, derived from domestic ores, be most economically and efficiently introduced into the bath?" Certainly not in the form of spiegel or any other manganese-silicon alloy, since the use of either in this manner implies additional melting costs and, further, the silicon of the manganese-silicon alloy used in the basic bath becomes an added burden to a process already overloaded with this element. "By the use of these ores direct in the open-hearth furnace?" Perhaps, although here one is also limited by the silica content, which, if excessive, will unduly increase slag volumes and consequently be followed by higher iron and manganese losses. Apparently the alternative lies only in the use of these ores by the blast furnaces, and a production therefrom of irons carrying a high percentage of manganese (2.50 to 3.00) accompanied by silicons of 0.50 to 0.75 per cent. and a maximum phosphorus of 0.400 per cent. This latter maximum is a purely arbitrary assumption, since no opportunity has presented itself to the writer for observations of the characteristics of this element above 0.350 per cent. It is apparent that this plan of operation must show manganese recoveries commercially comparable with present practice, and commensurate with the necessity or desirability of utilizing these ores. A comparison of this nature must not be confined to the pounds of manganese per ton of ingots produced in in the open hearth, but must also reflect the enhanced profit to the plant as a whole, which undoubtedly will come from lower rejections and decreased chipping costs, to which must be added that intangible important asset, good will, which comes only from a satisfied clientele of customers.

Exactly what may be expected in the basic open-hearth plants operating with an iron of this nature is difficult to state positively without the accumulation of a large amount of quantitative data. However,

experience at one plant working a 1.90 per cent. manganese iron has indicated fairly economical manganese recoveries, combined with very marked decreases in rejections and sulfur content of steel. It is pertinent to add at this point that 95 per cent. of all heats investigated for residual manganese carried a 5 per cent. initial ore charge, the remaining 5 per cent. ranging from 9 to 12 per cent. ore; that neither minimum nor maximum residuals appear in the high-percentage ore column; and that one 12 per cent. ore heat shows a constant residual of 0.31 per cent., which is four points above the average of 0.27 per cent. From this, it seems fair to assume that, with a certain degree of manipulative care, it is not impossible to obtain reasonably fair recoveries of metallic oxides from primary slags before wasting them from furnaces.

Inspection of data derived from twenty-five heats made at different periods, without in any great degree departing from standard practice, reveals the following facts: (1) That residual manganese at 0.30 per cent. predominates throughout a period extending from complete melt to tap. (2) That 0.35 per cent. has been attained in three cases, and in only one has it fallen to 0.19 per cent., with an average of 0.27 per cent. (3) That generally with FeO and MnO in the slag, approximating 15 and 12 per cent. respectively, the average residuals prevailed, provided the CaO content did not fall below 44 or 45 per cent. (4) That the high residuals were almost invariably the accompaniment of a CaO content of 47 to 50 per cent., with no sharply marked increase of the MnO, but did apparently bear some indirect relation to a proper degree of slag fluidity. (5) That these residuals persisted at carbons as low as 0.04 per cent. (6) That there was a very marked decrease in the sulfur content of the steel and also in the finishing-mill rejections.

These heats were made under conditions far from ideal for the most efficient manganese recovery, for, due to siliceous fluxes and refractories, plus high-silicon iron (1.25 per cent.), slag volumes were in excess of what the best practice along these lines demands; and due to a low grade of gas coal, temperature control was practically absent.

From the viewpoint of an exact scientific investigation, the data, on which is based this method of utilizing these low-grade domestic ores, may appear somewhat fragmentary, yet the conclusions drawn are substantiated by results attained by many others who have worked on a large scale along identical lines. So I feel little or no hesitancy in saying that given a high-manganese iron (of previously outlined composition) combined with the lowest possible extraneous silica from fluxes and refractories and a fuller appreciation on the part of the open-hearth personnel of the importance of slag manipulation and control, residual manganese may be driven to and held at 0.40 per cent., that the additions of high-percentage manganese alloys from this point on will be mainly a function of chemical specifications, plus the necessity of making up some

small volatilization losses that occur when tapping, and that these additions need be only in such amounts that there will be found not only a material economy in manganese in the high percentage alloys derived from foreign ores, but a reasonable recovery from the domestic ores and the certainty of a more perfect final product.

J. R. CAIN,* Washington, D. C. (written discussion†).—This paper is a contribution to a field of investigation that has been somewhat neglected; namely, that important stage of steel manufacture known as deoxidation. Its paper deals with two particular forms of manganese for deoxidation and two combinations of manganese and silicon for the same purpose and shows how, even with the use of comparatively few deoxidizing substances, or combinations, many procedures can be developed. The deoxidizing metals known to be useful for this purpose in the steel industry are: silicon, aluminum, manganese, titanium, and, to some extent, magnesium and vanadium. If these are combined two at a time, a new set of deoxidizers is available; if ternary or quaternary alloys are made, other sets with new properties are obtained. If such deoxidizers are applied with some of the variations in practice frequently used (such as in the quantity used, premelting of the deoxidizer, addition in the ladle or in the furnace or both, treatment in stages using a different kind of deoxidizer at each stage, etc.) it can be seen that the steel maker has available a great number of deoxidation processes, many of which have never been tried. Moreover, there is a strong probability that manganese, silicon, aluminum, titanium, magnesium, and vanadium are not the only elements available for deoxidation. Some yet untried may be even better for the purpose. Quite likely, too, valuable properties can be imparted to the steel by new variations in deoxidation practice.

The war has emphasized our military need for a systematic study of deoxidation to save manganese, of which the country lacks a sufficient amount within itself for the steel industry. Any interruption during war to importation of manganese ore would be fatal if processes were not at hand for using less manganese, lower percentage ferro, or substitutes. Work along these lines was undertaken during the war by the Bureau of Mines, Bureau of Standards, and National Research Council, in coöperation with steel companies, and it is hoped that this work may be continued, particularly on the broader lines of a scientific study of deoxidation and of new deoxidizers.

Viewing the subject historically, it is evident that the adoption of the deoxidizing agents at present in use was the result of practice developed more or less without scientific knowledge of the mode of action of the deoxidizer. Even now the exact chemical changes affected by the deoxidizer are not known. In fact, the very name "deoxidizer"

* U. S. Bureau of Standards. † Received Feb. 18, 1919.

indicates lack of exact knowledge. Actually, as will be shown by some papers soon to be published by the Bureau of Standards, the deoxidizer exerts but little influence on the oxygen content of a steel—at least in so far as this is determined by some of the frequently used methods. Part of the trouble lies in these methods, as will also be shown by future Bureau papers, but the important point is that the deoxidizing alloy is really added to remove the spongy structure of the metal due to gases. The exact nature of the gases and how they are affected by the deoxidizer has never been satisfactorily determined in a thorough and convincing manner, although numerous workers have thrown much light on various phases of the subject.

The Bureau of Standards is now attempting to develop methods for determining the gases in steel that are affected by the deoxidizer. These methods will then be applied to the analysis of steels in various stages of their manufacture, so that eventually some idea may be gained as to the relative efficiency of different commercial methods of deoxidation in accomplishing what they are intended to do; namely, removing or eliminating the effect of gaseous impurities in steel. Other work is under way with the idea of trying new deoxidizers and deoxidation methods on small heats of steel made under experimental conditions. The progress of the deoxidation will be followed by gas analysis of the metal, and the physical properties of the steels will be fully determined after forging, rolling, and heat-treating as commercially practised.

The foundation for intelligent coöperation in this work between various governmental agencies has been laid during the war, and it is hoped that this coöperation may continue along with an increasing degree of interest on the part of steel manufacturers, who will be aided in their work by every step in the investigations.

Water-cooled Equipment for Open-hearth Steel Furnaces

BY WM. C. COFFIN,* PITTSBURGH, PA.

(New York Meeting, February, 1919)

THE refractory linings of open-hearth steel furnaces above the bath line are subject to severe wear not only from the heat caused by the combustion of the fuel and the reactions of the bath, but also from the melting action of the gases when not controlled and directed, and from the scouring of the oxides, etc. with which the products of combustion are charged.

The framework that encases or binds the furnace is made up primarily of an iron or steel pan in which the furnace proper is built, side and end buckstays, and tie-rods. The expansion and contraction are so great and variable that the bindings may be subjected to severe stresses unless the tie-rods are carefully adjusted in heating up the furnace, but they are most affected by direct heat and usually become deformed from that cause when the lining is worn thin and, sometimes, when sections of the lining fall in and expose some of the framework.

A deformed framework displaces the skewback of the roof arch and tends to distort or crush the roof. Bent buckstays deform the front wall, causing the door frames to stand away from the lining so that the gases burn between the lining and the frames, destroying both of them. Frequently the frames are attached to the buckstays; then when the latter bend the frames usually break.

Preventing the distortion of the framework and maintaining the economy of the refractories are real factors in the cost of steel making. Not only does the cost of general repairs and reconstruction add directly to the cost of steel produced, but the time lost in making repairs cuts down the steel production and increases the overhead expense of both the furnace shop and its allied foundry or mills.

Many years ago water-cooled equipment applied to iron blast furnaces made it possible to increase greatly not only the size and production of the furnaces but also the life of the lining. It also cut down the loss from frequent repairs. The open-hearth furnace is a much later invention, especially as a commercial factor; for though it dates from the Fred Siemens patent of 1856, much more progress was made in the Bessemer process until the latter eighties and most of the development

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of open-hearth furnaces, especially in large furnaces and aiming for big production, dates from the memory of men still "in the game" and does not go back more than 25 years. Most of the advancement in water-cooled equipment is of even later date and it is within the last 10 years that it has become quite general practice to apply some form of water-cooled device to protect and increase the life of the ports and bulkheads, which suffer most from wear and tear and which require the most attention and expense to maintain.

Cast-iron and cast-steel doors and frames were water-cooled by means of hollow castings in which water circulated and by casting pipe coils in them. These are still used at some plants, but a casting is not a good enough conductor for the water to sufficiently cool the outside surface to protect it, so it is liable to burn out and also to crack and warp. As the heat is very much greater in the ports, other means are necessary to protect them. Quite a few copper-bronze water-cooled castings have been used with some success, but they can be successfully made only in moderate-sized castings and are easily pitted and injured by slag.

Adaptability of Steel Plates for Water-cooled Equipment

Steel plates about $\frac{3}{8}$ in. (9.5 mm.) thick can be readily formed into almost any shape desired for water-cooled equipment for doors and frames, buckstays, skewbacks, port coolers, division-wall coolers, bulkhead coolers, etc. If properly designed and flanged to keep the joints from undue stresses, the joints can be welded. Riveting is undesirable both on account of the liability of leaking and because the doubling up of thickness at the joints makes proper cooling impracticable. Steel plates have all the qualities of strength, ductility, and conductivity that make them especially applicable to this use, and they can stand contact with furnace slags and oxides without injury. They are also more free from defects and can be more readily and completely inspected than other materials. The conductivity of steel plates keeps the outside surfaces at nearly the temperature of the water in the device so that if it is in contact with the refractories it has much the same action as coolers in blast furnaces.

In contact with brickwork, which is a very poor heat conductor, the cooling effect will penetrate, at most, about 6 in. (152 mm.) and usually be effective in keeping the brick from softening with the heat for only 4 in. The result is that this type of equipment, if covered a few inches deep, does not have any cooling action on the interior of the furnace and, therefore, does not add to the fuel cost nor cause the furnace to be sluggish. This is at once noticeable with the $4\frac{1}{2}$-in. brick lining on steel-plate furnace doors which forms a slag coating

on the exposed face after thinning down 1 in. or less and will frequently last a full campaign of the furnace run unless jarred out.

About the year 1900, L. L. Knox, of Pittsburgh, Pa., who was engaged in the construction of open-hearth furnaces, began the study of water-cooled equipment and has since devoted most of his time to the development of these devices. With the exception of a few riveted steel-plate doors and frames, he at first designed and installed copper coolers. In 1909, he turned his attention entirely to welded steel-plate devices, which have come into such general use that the latest developments should prove of interest to all engineers in any way associated with the open-hearth steel industry, or who wish to keep up with the growth of its equipment. Many of these devices have a much wider application and are already installed on heating furnaces, soaking pits, copper furnaces, and glass-melting furnaces, and engineers familiar with these various types of furnaces can see where they can be applied to increase the life and reduce the maintenance cost. These devices have been patented by Mr. Knox, with the exception of the Neeland buckstay, which was patented by M. A. Neeland, formerly chief engineer of the United States Steel Corp. They are manufactured and sold outright without any royalties or limitations as to their use.

Illustrations of Water-cooled Equipment

Fig. 1 illustrates an open-hearth steel furnace equipped with water-cooled devices at all essential points, each of which will be referred to separately.

Fig. 2 shows a door and frame, all welded joints of which are kept away from both the back of the frame and the door opening. The back plate is flanged through the door opening and turned back about 2½ in. onto the front plate, which not only takes the welded joint away from the extreme heat but strengthens the door arch, which is subject to stresses from expansion and contraction. The stay-bars and stay-bolts do not penetrate the back plates of either doors or frames, but are welded onto the inside surface before the outside plate is welded; this avoids any possible leaks against the brickwork. Many of these doors and frames have been in continuous use for 3 or 4 years, whereas cast ones seldom last more than a few months. The jambs and arches also last much longer when in contact with these frames and the furnace operators can give the process of melting more efficient attention on account of the cool front.

Fig. 3 shows the doors and frames connected with a direct feed-water line. The raising and lowering of the door is taken care of by a telescopic inlet pipe.

Fig. 4 shows a circulation water system with but one feed-water

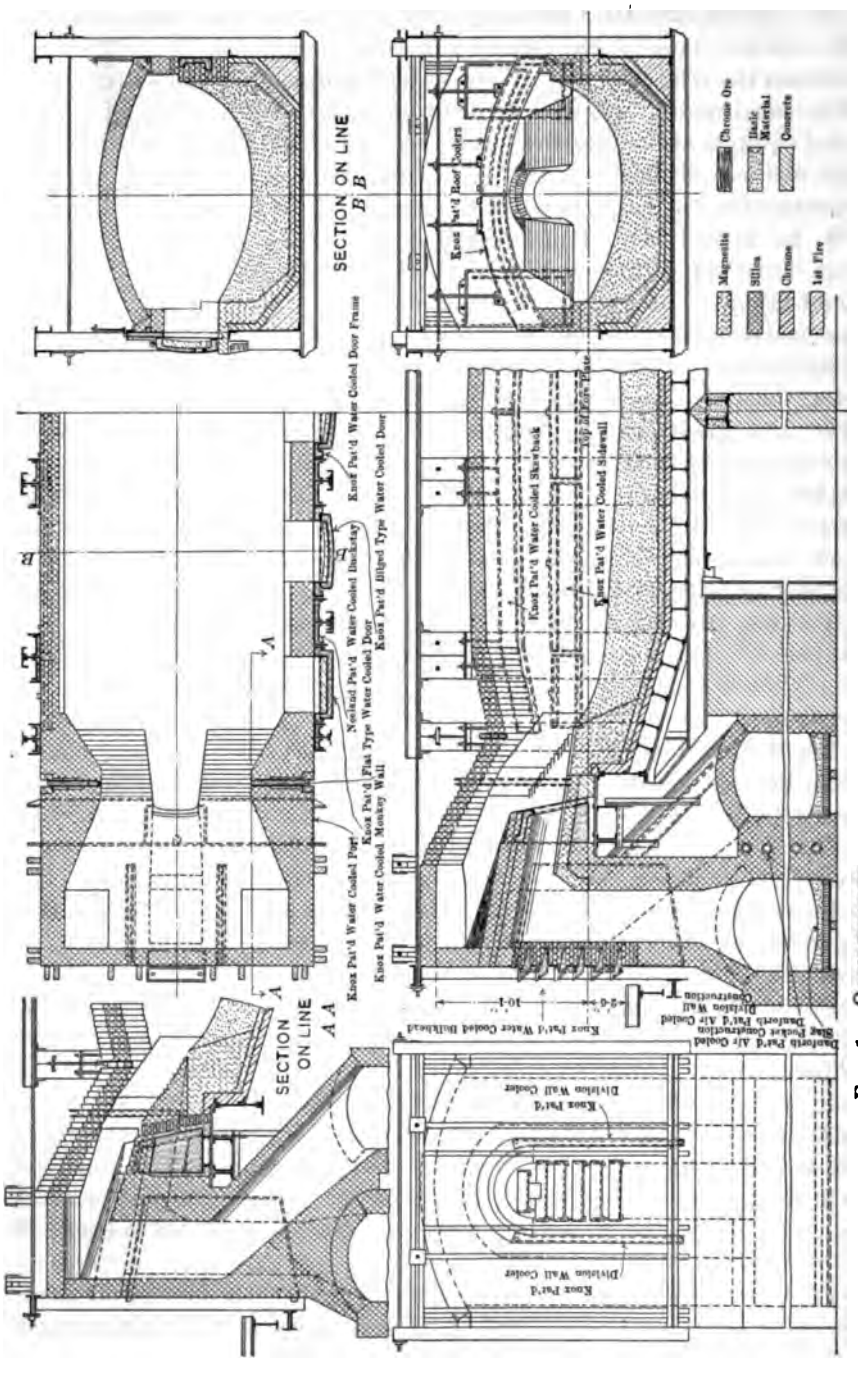

FIG. 1.—OPEN-HEARTH STEEL FURNACE, EQUIPPED WITH WATER-COOLED DEVICES.

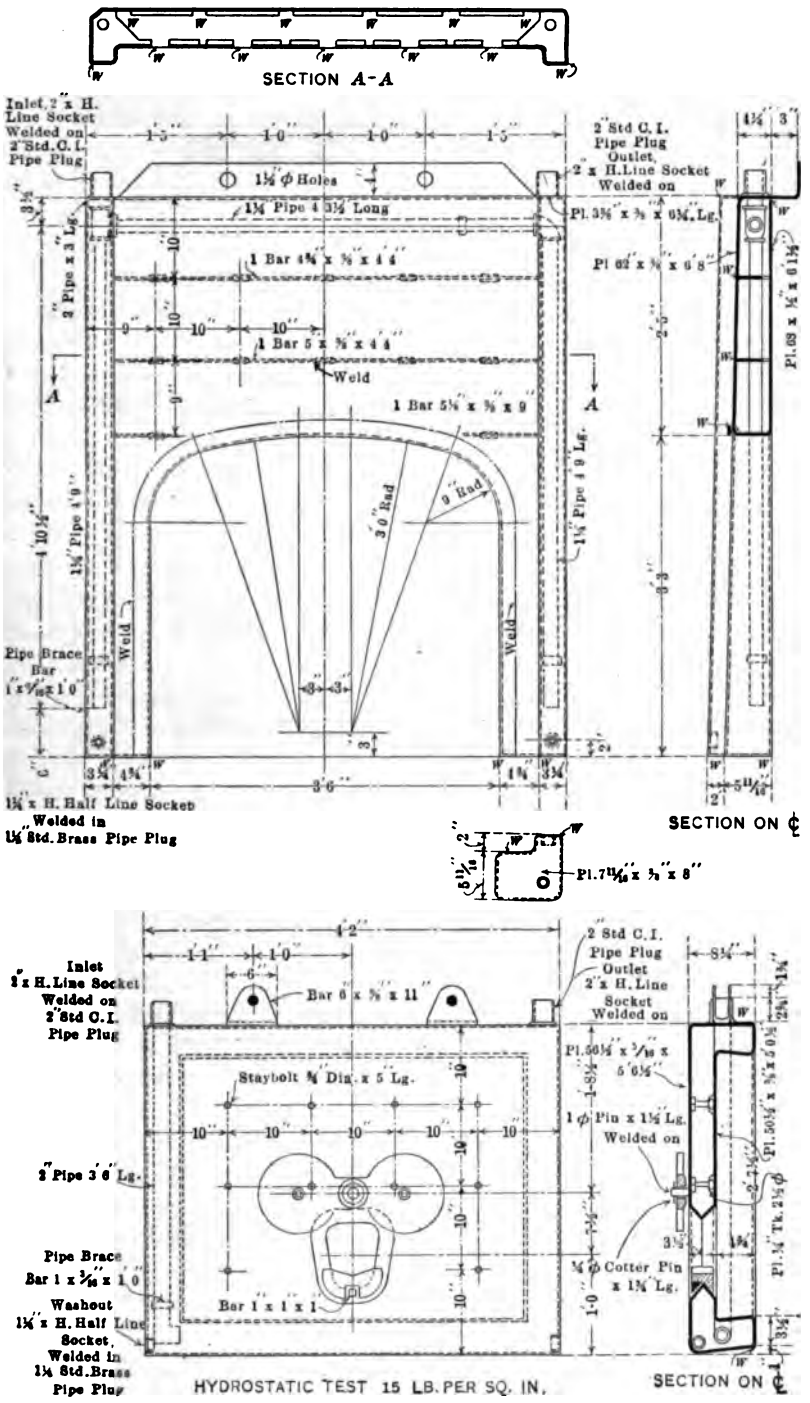

FIG. 2.—KNOX PATENTED WATER-COOLED DOOR AND FRAME.

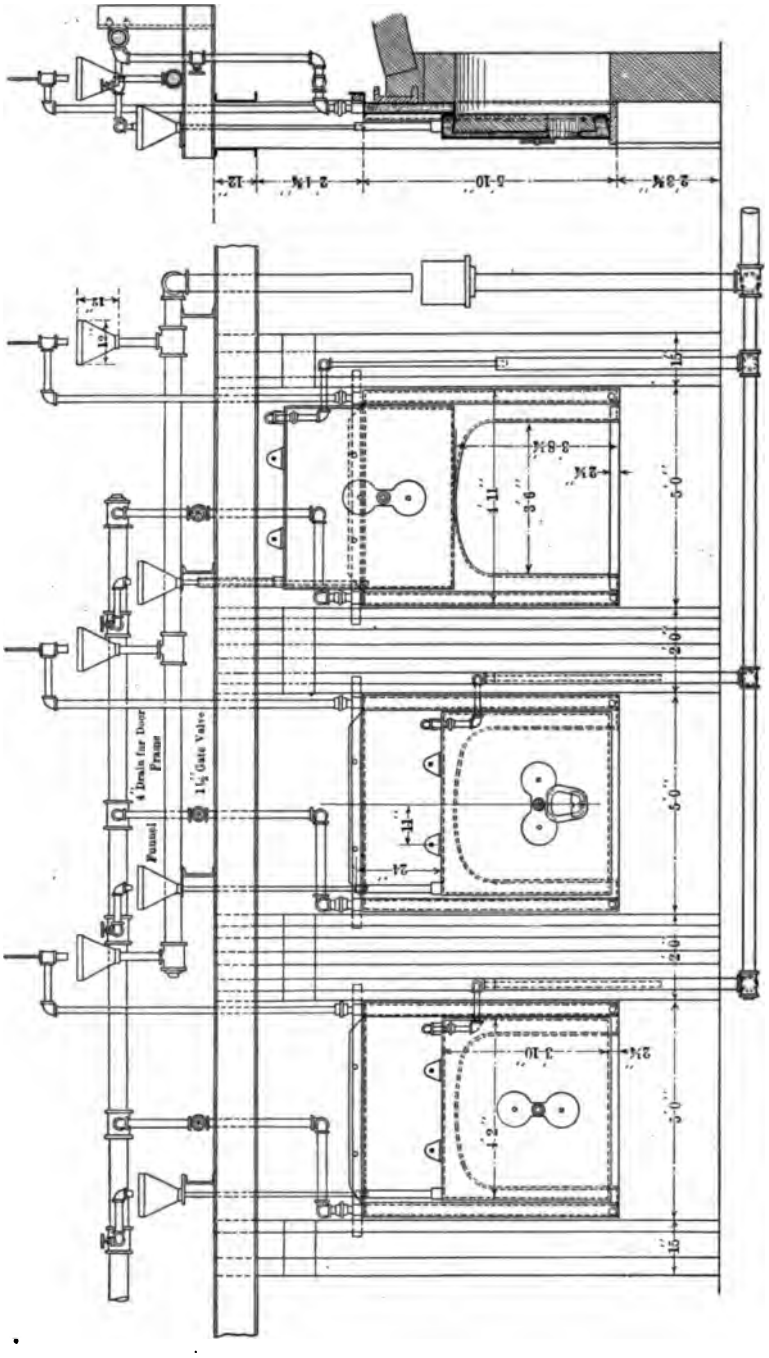

FIG. 3.—DIRECT FEED-WATER CONNECTIONS FOR DOORS AND FRAMES.

connection for the furnace front. This feed line has a nozzle entering a cast-iron ejector on the bottom of the circulation tank, connected with a manifold for feeding the door and frames. The doors are connected with a hose both for inlet and outlet, and the inlet water from both doors and frames is returned to the tank. The overflow from the tank is just sufficient to keep the water from boiling and is usually held at a temperature of about 180°. By having the overflow pipe enter a waste line through an open funnel, the amount of water can always be regulated to suit the requirements by merely inspecting the overflow and seeing that the temperature is neither too high nor too low. The circulation system reduces the amount of water about 30 per cent. below what is used with direct feed, and the circulation tank acts as a reservoir in case the feed water is cut off by accident until either the feed line is repaired or an emergency connection is made with the tank.

Figs. 5 and 6 show a circulation system applied to a tilting furnace, taking care also of chills at the connection between the furnace and the port ends, and of port coolers. Fig. 5 shows the main furnace and Fig. 6 the port end, each with its separate circulation system.

Fig. 7 is a detail of the Neeland water-cooled buckstay, which is also shown in Fig. 1. When used in connection with water-cooled doors and frames, it protects the entire front of the furnace. These buckstays will always keep in alignment; that is, they will not bend on account of heat and, consequently, they hold the skewback of the roof arch in its proper place, with the result that not only is the front wall sufficiently cooled to hold it in place long after it would be burned out with a dry buckstay but the skewback is held in its proper place and relieves the roof from being crushed or distorted.

Fig. 8 shows the port end with the location of the port cooler. This cooler is carried across the port end the full width and both ends of the cooler are exposed; this gives access to the water connections and allows the cooler to be put in place and removed by the charging machine. The charging machine, of course, brings it out only far enough for the overhead traveling crane to get hold of it; when there is no traveling crane, the cooler is placed half way into the furnace with block and tackle and the charging machine pushes it into final position. The port cooler is located at the point where the mixture of the gas and air creates the best combustion, the gas being directed toward the bath. With a dry port, the brickwork is built much nearer the bath than it should be but this is necessary to allow for its wearing away, and it is not rebuilt until it is worn back beyond the point of good combustion. This construction not only makes a waste of fuel, but the lack of control of the direction of the gas causes it to burn both the roof and the side walls. This burning is prevented by the port cooler, as the gas port is held to its proper position and also to its original size, with the result that the furnace works sharply

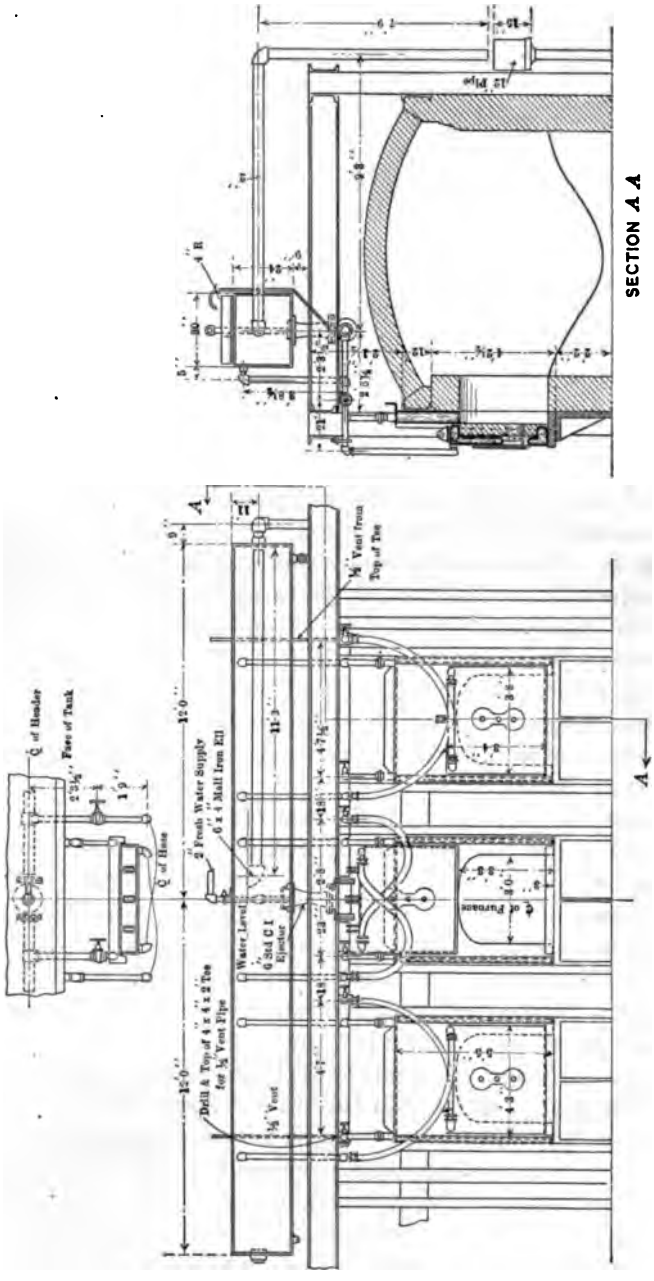

FIG. 4.—CIRCULATION WATER SYSTEM.

and continuously until the checkers become so clogged that the furnace must be shut down to clean them and rebuild at least the top courses.

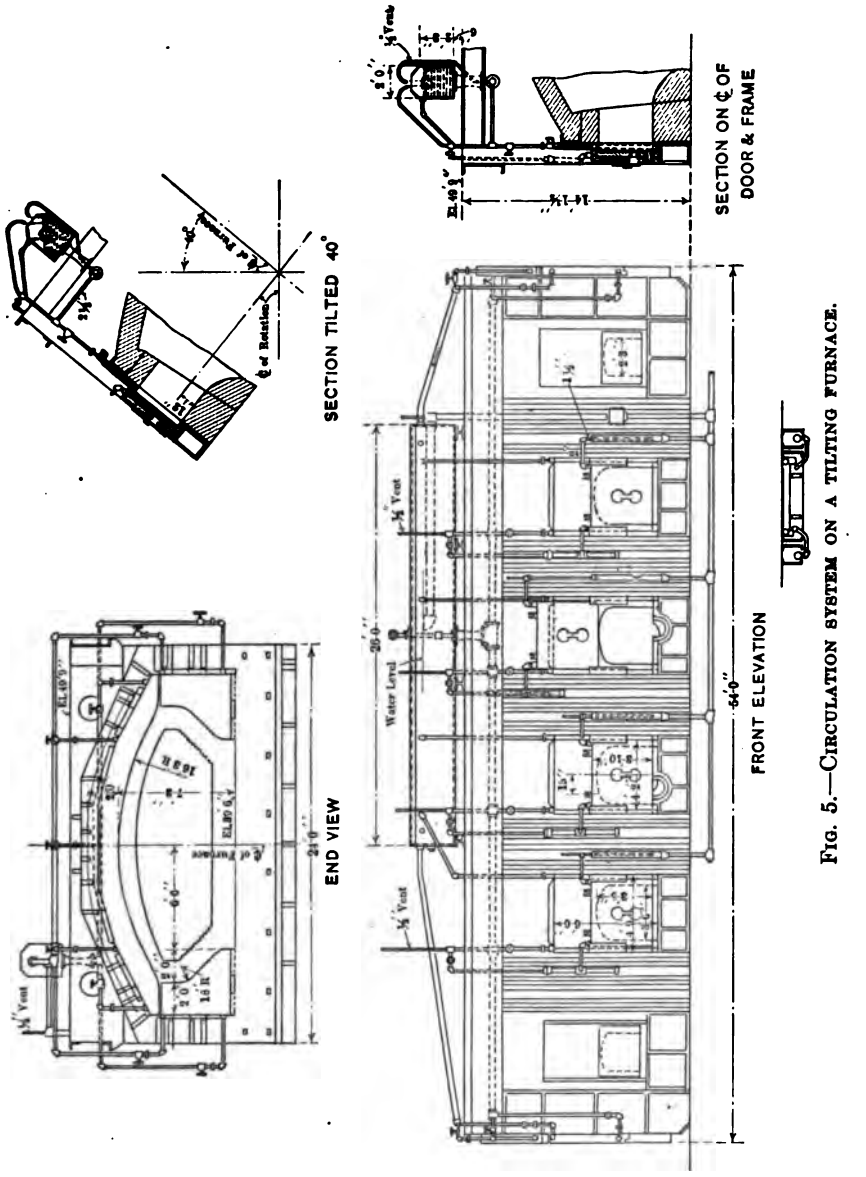

FIG. 5.—CIRCULATION SYSTEM ON A TILTING FURNACE.

Fig. 9 shows a detail of the Knox port cooler, which is so constructed that a 4½ in. (114 mm.) brick lining is built in the gas port in a recess, so that this lining is protected from being destroyed by the products of

combustion. The outlet water is taken by an internal pipe from the highest point of the port so as to use the minimum amount of cooling water and make a steam pocket impossible. The port coolers work best

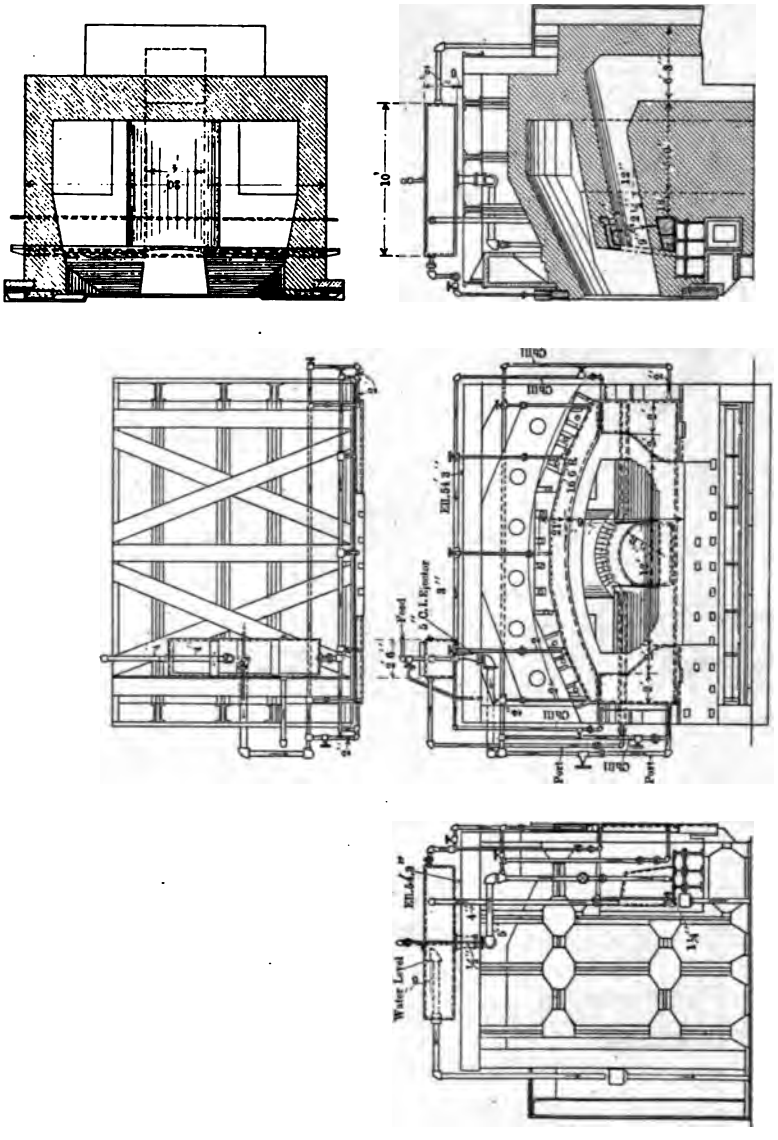

FIG. 6.—CIRCULATION WATER SYSTEM APPLIED TO THE PORT END OF A TILTING FURNACE.

with a small head of water, for which reason the outlet pipe is carried up about 6 ft. with a vent on top, so that there can be no siphoning. These pipe connections are shown in Fig. 10. The port cooler has a large chamber in which sediment can settle without any detriment to the

cooler, or the working of the furnace. In many cases the coolers, when cleaned during the reconstruction of the furnace or checkers, have been found to contain 8 or 10 in. of mud or coal dust, or whatever impurities

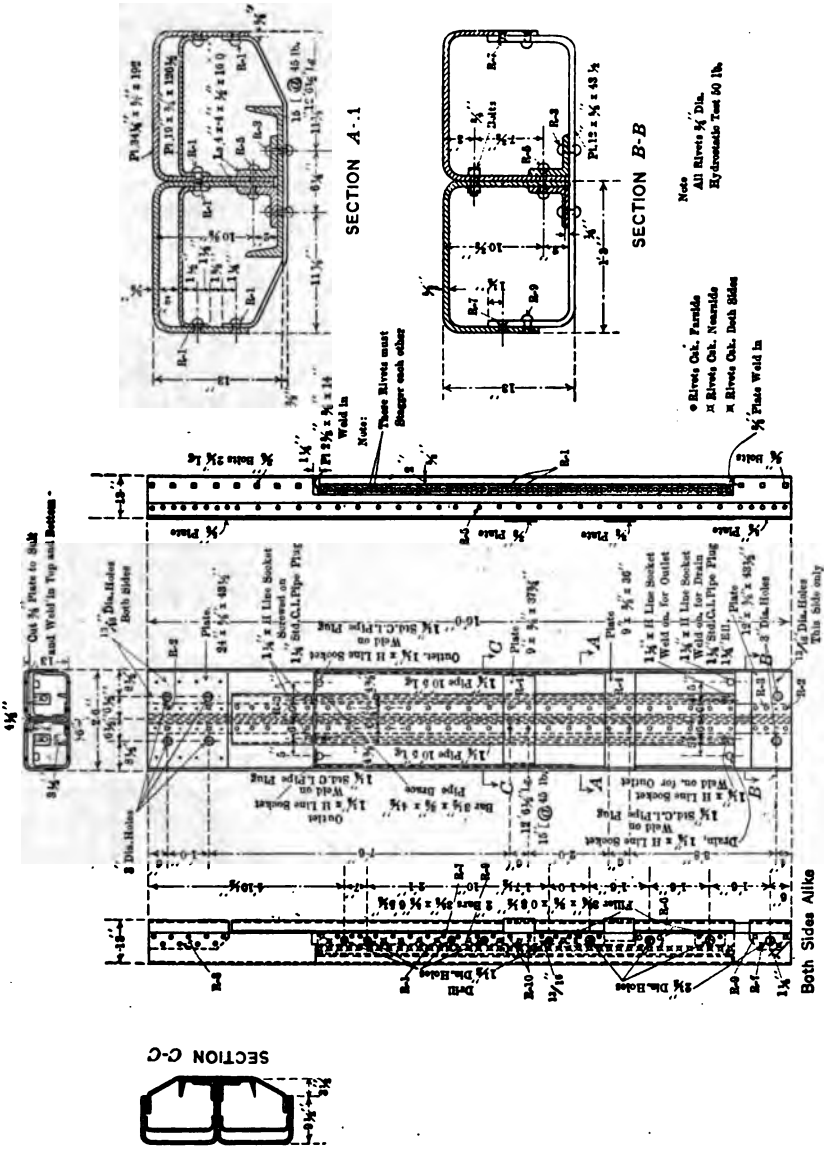

FIG. 7.—NEELAND PATENTED BUCKSTAY.

the feed water carries. Many of these port coolers have been in use for several years, and have a record of from 1500 to 2000 heats. The design has been improved from time to time and indications are that they will give better service as now constructed.

Fig. 10 shows the coolers in the bulkhead that forms the back wall of the gas uptake. This is subject to a great deal of wear, as most of the heavy oxides and other solids carried over with the products of combustion go

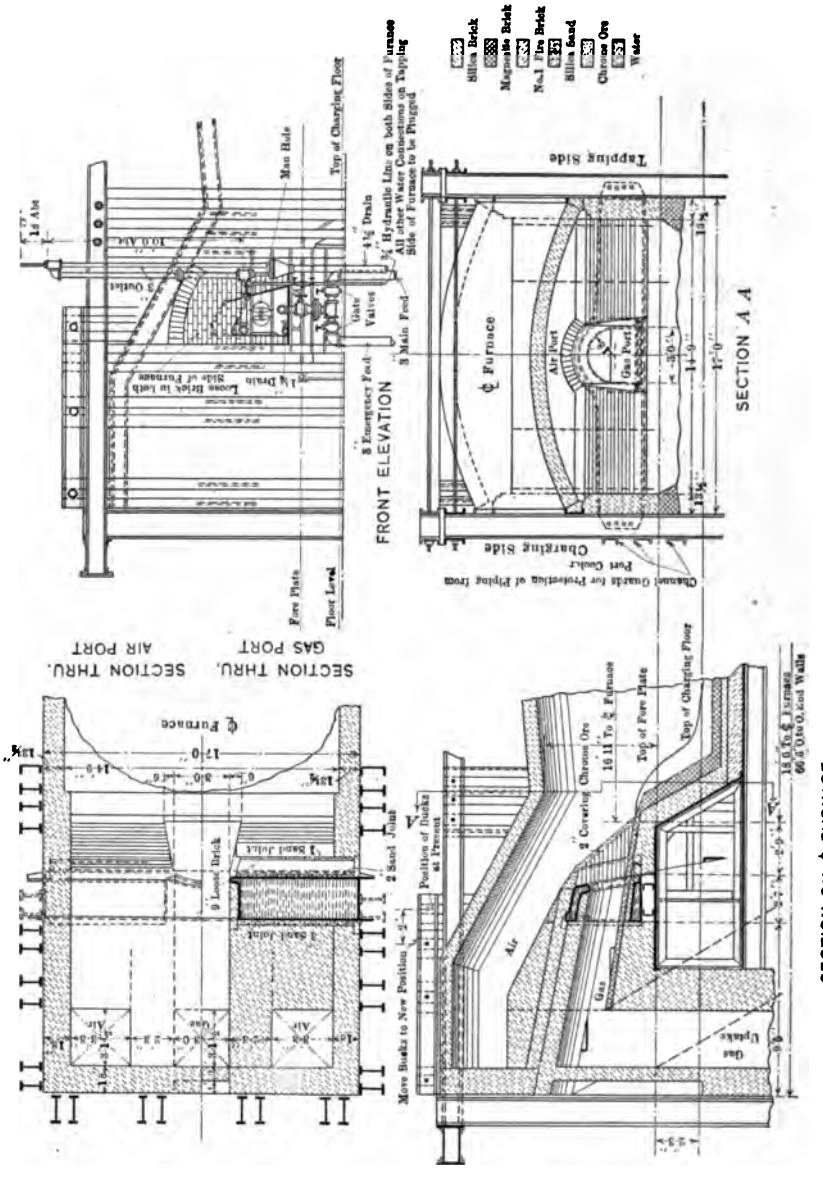

FIG. 8.—PORT END SHOWING LOCATION OF PORT COOLER.

through the gas port and impinge on the end wall of the gas uptake. The bulkhead coolers are arranged in a series one over the other, somewhat like bosh plates in a blast furnace. There is a tendency for the

division walls between the gas uptake and the air uptakes to either crack or burn out so that the gas and air mix, causing premature combustion. This is prevented by a vertical water-cooled slab or division wall cooler, also shown in Fig. 10.

In order to secure the best combustion in the furnace and also to protect the front and back walls, it is desirable to contract the opening by thickening up the side walls in front of the gas port. This heavy wall is called the monkey wall; frequently furnaces are built without monkey walls because they are destroyed so quickly by the products of combustion. Monkey-wall coolers, which are shown in Figs. 1 and 10, are water-cooled slabs about 2 ft. (0.6 m.) wide and 7 ft. (2 m.) high, which project through the roof. This permits them to be easily placed and removed and also gives access to the water connections. It also forms a seal between the monkey wall and the roof, where the monkey wall first crumbles when it is not protected.

Fig. 11 shows a general arrangement of the Knox gas- and air-reversing valves and Fig. 12 is a detail of this valve. The Knox valve goes back to the first principles, as it is based on a simple vertical damper operating directly in the flues. Other types of valves operate above the flues and require uptakes and downtakes, which reduce the draft usually to 40 or 50 per cent. of the stack draft. They have no means for regulating the draft except by a stack damper, which does not make any individual control of the furnace ports nor a separate control of the gas and air checkers. Each Knox valve is a regulating damper; and when most of the products of combustion go through the gas checkers, leaving the air checkers cold, the gas valve can be opened very slightly, or not at all, causing the draft to come through the air checker and heating it. The valves are slightly inclined so as to insure close contact and tightness between the damper and the seat. There was some question as to the effect of tar on the surface of the seat, but it has been found that the tar is only a lubricant, and insures tightness. The valve damper is constructed with a thin steel frame so that the contact between the damper and the seat is narrow, but both the damper and the seat are machined so as to insure flat surfaces. The construction of the damper, with corrugation on the plates, takes care of a slight difference in expansion and contraction between the plates and the outside frame.

These valves are now installed in about 75 furnaces and are a great aid in having the furnaces work sharply, on account of the assured draft and the individual regulation of each chamber. The operating rigging, which is in most cases worked by an electric motor, is so designed that the gas-reversing valves work in a series ahead of the air-reversing valves so that there is no mixture of gas and air, which might cause an explosion, especially when waste-heat boilers are used.

Fig 1 shows roof-arch coolers, back-wall coolers, and water-cooled

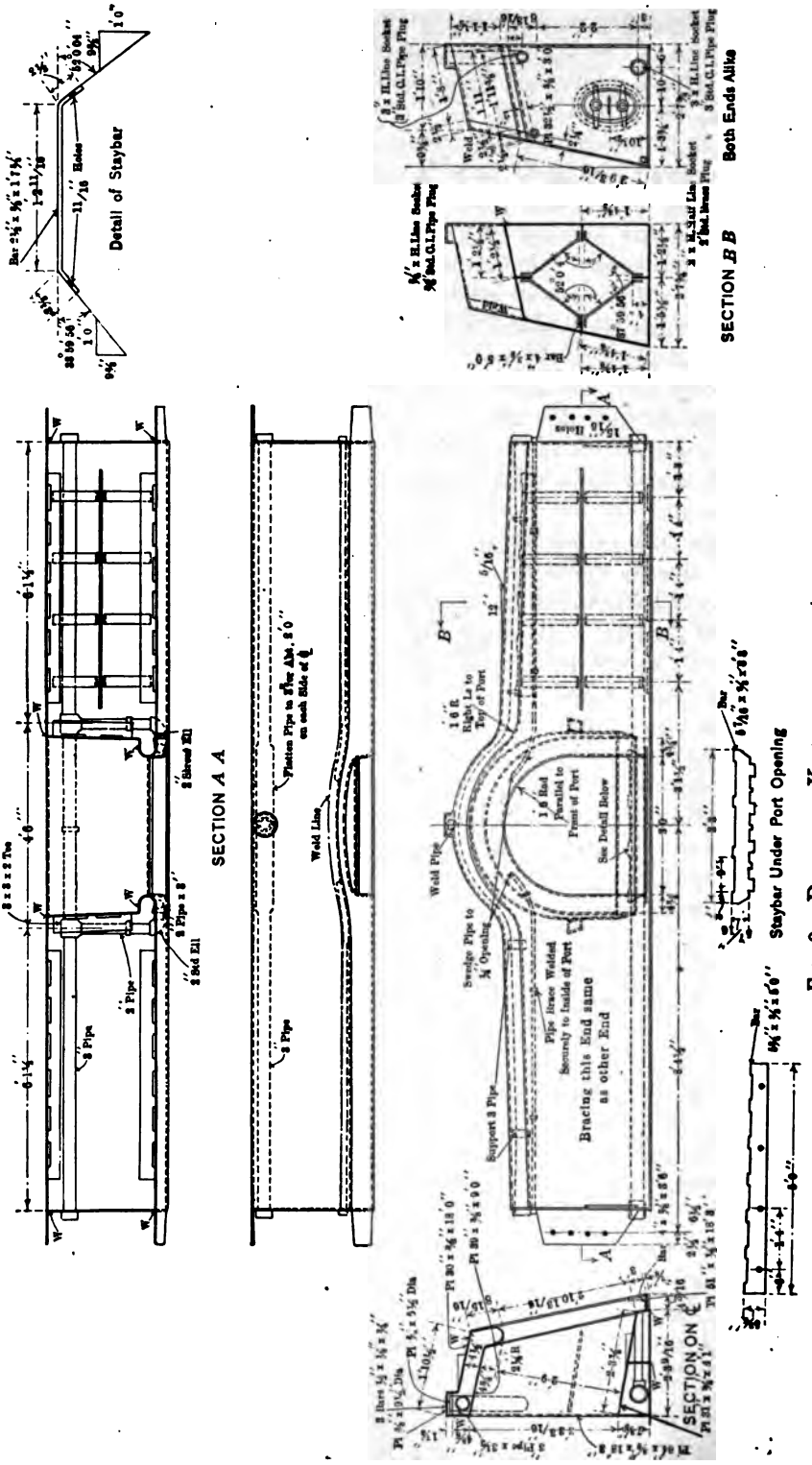

FIG. 9.—DETAIL OF KNOX PATENTED PORT COOLER.

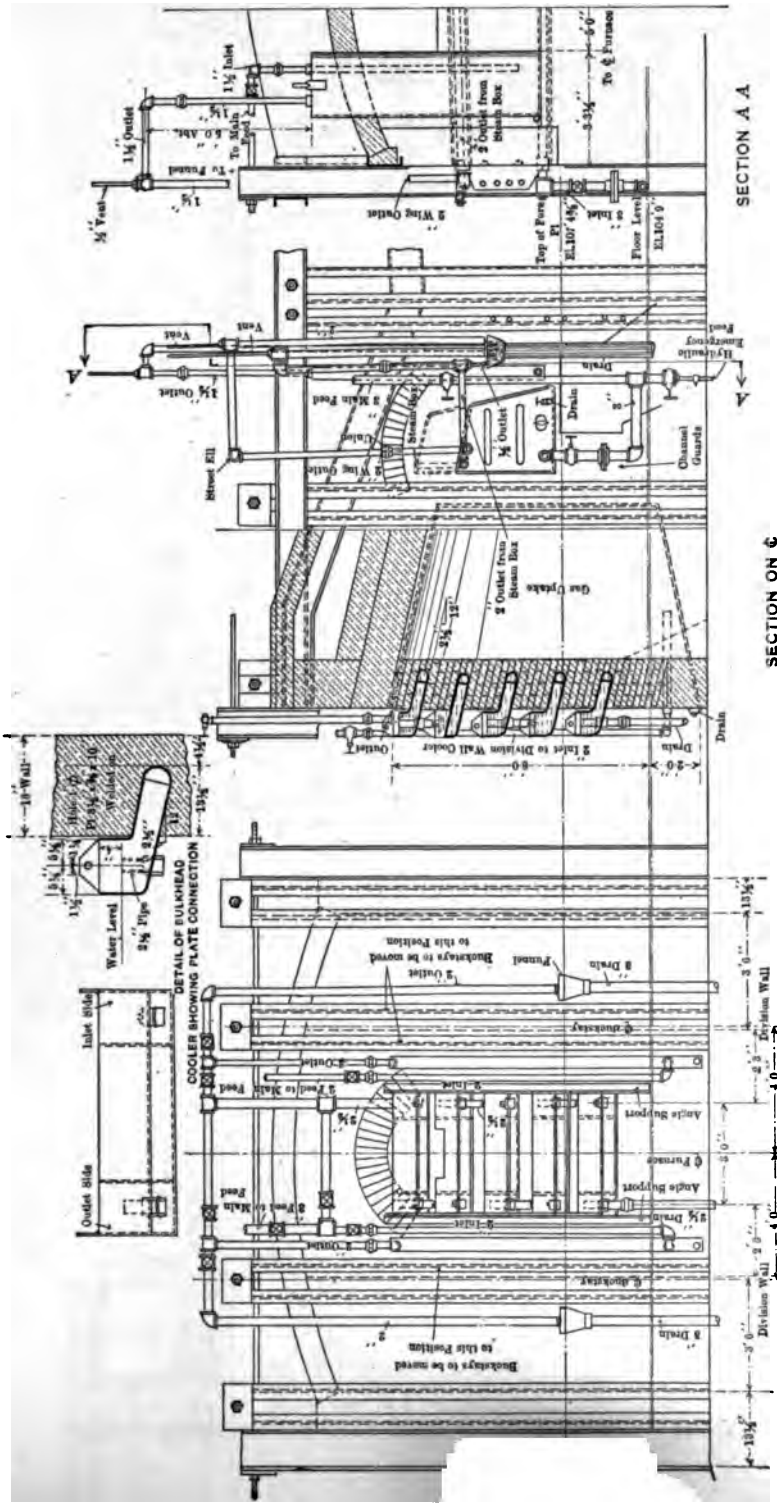

FIG. 10.—PORT END SHOWING BULKHEAD DIVISION-WALL AND MONKEY-WALL COOLERS.

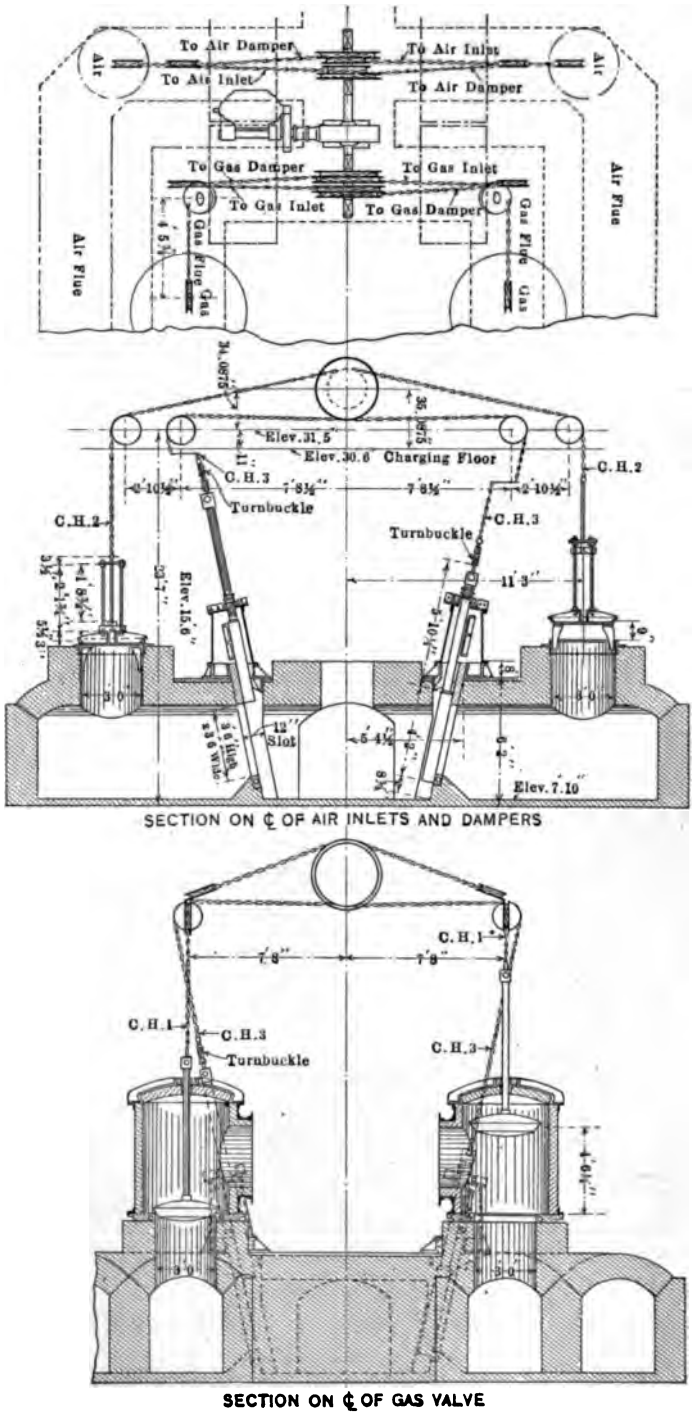

FIG. 11.—ARRANGEMENT OF KNOX PATENTED REVERSING VALVES.

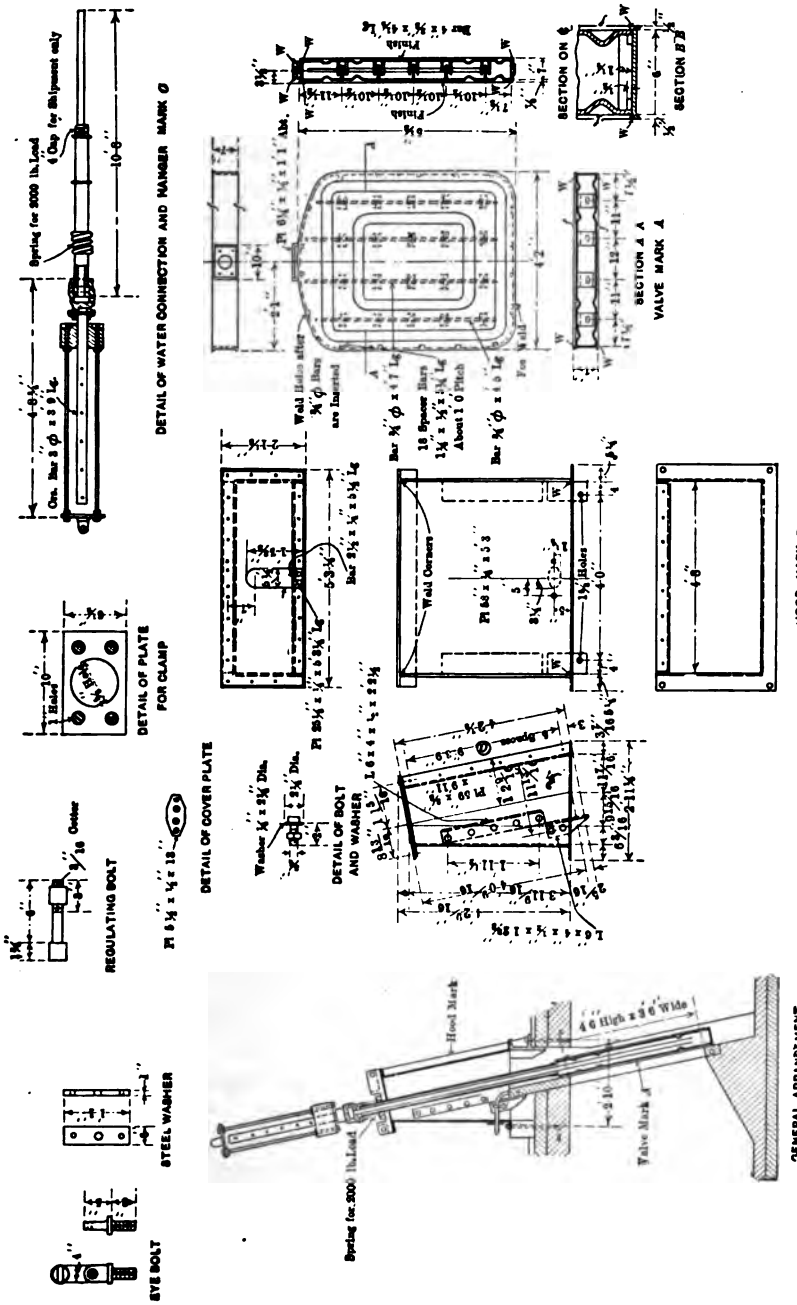

FIG. 12.—DETAIL OF KNOX VALVE.

skewbacks. These are not in very general use, but have their advantages wherever the operators find that their furnaces are cutting out at these different points.

Although there is a general tendency among managers of American plants to exchange information of value and operators from one plant are usually welcomed at other plants, direct reference to the installation of special apparatus should not be published. For this reason the writer has refrained from mentioning where the several water-cooling devices described are in use. Engineers interested in the general problem will find these devices in almost any plant they may visit and will also learn that the problems to be solved in one plant vary considerably from those in others. The only way to secure the best results is to consider each furnace shop as a different proposition, so that the cooling devices may not only be installed where they are most needed, but that the details are made to especially suit their requirements.

Amount of Circulation Water Necessary

The amount of water necessary for circulation will vary not only with the size of the different devices but with the condition of the furnace. It is dependent also on the quality of the water. When the water contains lime or other solids that precipitate when the water is heated, it should at once be determined in the laboratory at what temperature the solids precipitate, and the outlet water should be kept from 10° to 15° below this temperature. This may require using more water but insures a longer life for the equipment and less trouble for the operator in cleaning it. If left to the furnace operators, there is usually much water wasted, as their interest is in the steel making and they do not have the time to watch the equipment closely. It is better practice to have a pipe fitter look after the cooling devices to see that the outlet water is as hot as it can safely be used. When the installation is first made, he should give the apparatus very close attention for the first few weeks, starting with more water than is necessary and gradually cutting down the supply. By following this method it will be found that the minimum quantity of water will give the best results. Unless it contains a precipitate, there is no reason why the outlet water should not have a temperature of between 170° and 180°.

Conclusion

As stated at the beginning of this paper, water-cooling devices for open-hearth steel furnaces should in general follow the lines used in iron blast furnaces. They should not, in a practical sense, be exposed on the inside of the furnace nor tend to slow up operations but should only

absorb the heat that would otherwise be radiated in the air or be placed where they have a brick covering and keep it hard enough to hold and thus increase the life of the furnace and keep it true to its best lines. The writer has not had the time to incorporate in this paper the history of the development of these and other water-cooled devices; this will make an interesting subject for a paper in itself.

Many plant superintendents have devoted much thought to this subject and have developed equipment or contributed suggestions that have improved the devices herein illustrated and Mr. Knox can never speak highly enough of the help, coöperation, and encouragement that he has uniformly received from managers, superintendents, and engineers.

DISCUSSION

J. S. UNGER,* Pittsburgh, Pa. (written discussion†).—The author of the paper has given an excellent description of the appliances used to cool parts of an open-hearth furnace, and some of the reasons why the refractories fail in service. Personally, I favor water-cooling, but to a limited degree. At one time I advocated more universal cooling than is now practised, but believe I must change my mind. Water-cooled, thin-lined blast furnaces have not proved as satisfactory as anticipated. While this may retard the building of many such furnaces, it will not prevent the use of regular tuyere- and bosh-cooling appliances.

Water-cooling any furnace may be easily carried to the extreme. In an open-hearth furnace, excessive port or roof cooling, or cooling the working floors may preserve the lines of the furnace, lower the cost of refractories, add to the comfort of the workmen, and necessitate fewer shutdowns for repairs, but it may not be good economy on account of the high first cost, the cost of cooling water, maintenance of cooling appliances, and the immediate reflection in a higher fuel cost. The question becomes a financial one, and is answered by using water-cooling devices up to the point at which they cease to show true economies.

Open-hearth furnace bulkheads, roofs, and ports fail rapidly in about the order named. If they could be made to last indefinitely by water-cooling, the furnace would still have to be shut down periodically for cleaning the slag pockets, regenerator chambers, and flues. Extra, or spare, pockets and regenerator chambers have been advocated to permit of continuous operation, but the space required, the cost of installation, and the fuel required to initially heat such cold chambers will largely prevent their adoption. Some particular fuels, as high-ash powdered coal, require frequent stoppages for cleaning, but for this particular fuel a special provision can be made.

Water-cooled doors, frames, valves, and partly cooled ports are proper,

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and show economies, but it is doubtful if a completely cooled port, as at present developed, can be called a success. The difficulty of holding a refractory lining on the inner surface, the large volume of water required to cool the port, and the higher fuel consumption do not recommend it.

It is difficult to obtain a true comparison of the costs of a normal furnace with a full water-cooled port furnace, because the furnaces may not be operated under the same conditions as regards the charge, the fuel used, and whether the normal furnace was not water cooled or only partly cooled in the port. The only figures I have seen seem to show that the full water-cooled port requires a longer time to make the heat, more coal per ton, higher cost for cooling water, and a higher cost for the original installation; but it has the advantage of a longer life with lower refractory and rebuilding costs.

An adequate supply of water free from scale-forming salts is not always obtainable. When a heavy scale has formed, it may cause a part of the port to burn out, necessitating a shutdown and expensive repairs.

Water-cooling should be used on an open-hearth furnace at places where it can be protected by just enough thickness of brick to prevent high fuel consumption and where the cooling effect of the water is sufficient to prevent the brick from being rapidly destroyed; or, in other words, the proper thickness of brickwork to approximately balance the fuel cost as opposed to the refractory cost must be determined by careful experiments at each point in the furnace where water-cooling appliances are used.

L. L. KNOX,* Pittsburgh, Pa. (written discussion†).—This very interesting paper on water-cooling describes various devices for the different parts of the furnace, but overlooks the water-cooled chills for rolling open-hearth furnaces. These chills, shown in Figs. 5 and 6, make it possible to break the joint between the ports and furnace, allowing the furnace to rotate. They are, preferably, made in three pieces—one top and two sides. The top furnace chills must have three outlets to prevent steam pockets and two feed pipes; they consist of steel plates. The back plate is the length and height of the chill and is flanged in to suit the opening, and back about 2½ in. (6 cm.) on the front. The front plate is bent on straight lines, very close to circular. As many breaks are made in this as are necessary and it is welded to the flanged part and riveted at the top to the back plate. It has wearing strips made of cast or bent plates riveted above the cooling parts; these plates stand out about ½ in. (1 cm.) farther than the water-cooled part and take the rubbing, or wear, where the two chills sometimes come together.

As these chills and furnace parts can be made interchangeable, they

*Inventor of the water-cooled equipment. †Received Feb. 18, 1919.

are generally made of the same construction, with regard to inlets and outlets. The back plate of the side chill is flanged and bent back about 2½ in., forming the same construction as the top chill with the exception that a baffle plate is placed near the top to carry the outlet close to the inner corner to prevent a steam pocket. As these chills are flat and are cooled to the same degree all over and stay the same size throughout the run, the brickwork is cooled and protected and cannot cut away. This is impossible with castings water-cooled with pipes cast in them. The casting soon burns away and leaves nothing but the pipes, which are no protection to the brick.

Not many years ago, there was a firmly grounded prejudice against water-cooling any part of an open-hearth furnace, from the belief that water-cooling would rob the furnace of heat, which was considered the all-necessary requirement. Experience has proved, however, that the amount of heat, as represented by the heat units delivered to the furnace, is by no means the measure of the performance of the furnace. The design of the furnace lines, of course, has always been recognized as necessary to insure proper combustion at the proper place to insure the fastest possible working of the furnace, but the maintenance of these lines is also absolutely necessary. Since refractory materials will not withstand the destructive effects of temperature, abrasion, and chemical action, certain parts must be protected so that the furnace lines will not be seriously changed. The loss in furnace economy due to heat carried away in the cooling water is then much more than offset by the increase in economy due to the maintenance of proper combustion. Since water-cooling must be recognized as carrying heat away from the furnace, it is only logical that all cooling surfaces should be protected as fully as possible. As Mr. Coffin pointed out in his paper, the flow of heat through refractory brick is very slow. Compared to the total heat loss in a furnace system with low heat efficiency, the heat carried away by the coolers, when sufficiently protected, is not proportionately great. The one thing that must be kept in mind, however, is that while the furnace can afford to lose heat, without detriment to its operation, exposed cooling that appreciably affects the flame temperature will have a direct effect upon the furnace output.

The direct radiation from the hearth to the brick covering of any water-cooler is not great, since the temperature of the covering will never be more than a few degrees below the fusing point of the silica brick. When coolers are exposed, however, the radiation of the cold surface is very rapid and can seriously affect the furnace operation. Granting that water-cooling is a compromise between the sacrifice of heat economy and the maintenance of furnace lines to insure proper combustion, the question devolves into this: Is the loss of heat economy more than compensated by the gain in tonnage or the decrease in relining cost, or both?

The figures on which the comparison is based should cover sufficient time to have all factors developed. Furnace practice in this country is changing so that sharpness of the working of furnaces, with the consequent decrease in time making heats, is sought after more than length of furnace runs. The two things are somewhat opposed to each other. It is not to be expected, for instance, that a furnace that will make heats in 8 hr. will not suffer more abuse to the refractory lining than another furnace, having identically the same practice, that makes heats in 10 hr., due to not working as sharp; but the increased maintenance is much more than offset by the increased tonnage. I do not mean that the relining cost per ton of steel need necessarily increase in sharper working furnaces. Sometimes, a change for the better in certain lines of a furnace will cause more severe abuse on furnace parts that heretofore have given no trouble, but it only becomes necessary then to strengthen the weaker link in the chain. For example, if a furnace works continuously hotter due to more perfect combustion resulting from proper design and maintenance of ports, the uptakes on the outgoing end may give more trouble from burning through than previously. The corrections then should be applied, not in departing from the better design of ports but in preventing, by water-cooling, the burning through of the division walls. As the weak links in the chain become strengthened, heretofore strong links will become the weak ones. And let me say here, that in a division wall with heat on both sides water-cooling will help radiation by carrying the heat away rapidly and have no serious effect upon the working of the furnace.

All water-cooling devices should absolutely control the circulation of the water and have the outlet taken from the highest point. Where the top surface of the cooler is flat, or approximately flat, the exhaust should be carried, if possible, high enough to give 4 to 5 lb. pressure in the cooler. This can be done by carrying up the pipe, putting on a vent, and then carrying down the exhaust. Provision should also be made to wash out the mud, or have a large place for its deposition.

Water-cooling devices can be roughly divided into two classes: those designed to maintain proper furnace lines in order to insure proper combustion and faster working of furnaces and those that are used primarily to protect parts of the furnace not directly connected with the combustion of gases, and the use of which means an economy in repairs and maintenance and the saving of time due to elimination of many short furnace delays. In the first class are ports and monkey-wall coolers; in the second class, water-cooled buckstays, doors, frames, etc.

It is gratifying to feel that the firmly grounded fears of the early days that water-cooling would slow up the furnace have very gradually disappeared, as the experience of most of the plants in the country at the present time has shown them not to be well grounded. In my opinion,

the use of water-cooling in melting furnaces of all types is still much nearer its minimum than its maximum and a constant growth will continue, as in the past, which will result in increasing output and decreasing costs.

HENRY D. HIBBARD, Plainfield, N. J. (written discussion*).—Water-cooled devices for open-hearth steel furnaces are an American development and largely invented by Mr. Coffin's associates. As a new class of equipment, it would seem that they deserve a distinctive name and when the question of this paper came up it was suggested to Mr. Coffin that the abbreviated word "cood" (plural coods) be adopted; but no one-word name has been given them. These devices are used for two different purposes: lengthening runs between repairs and lessening the discomfort of the furnacemen. Some of them have undoubtedly come to stay.

Methods of Disintegration.—In considering the need of prolonging the life of the furnace parts, the different processes of disintegration or failure should be kept in mind. The roof, though it takes up some oxide of iron, which discolors the lower end of each brick, fails by direct fusion or disintegration because the hearth temperature is higher than it will stand. Similarly, the back wall and roof near it may be melted because the furnace is too hot on the back side, due usually to the improper guiding of the flame, while the front is cooler. This is a common fault in steel furnaces.

The faces of the ports, on the other hand, are wasted chiefly or wholly by the fluxing action of the basic oxide deposited by the outgoing gases, which action continues until the fusion point of the brick surface is low enough to cause it to melt, run down, and enter the bath. This decrease in heat-resisting power will go on until a composition is reached which gives a fusion point so low that the material will melt even when practically in contact with a water-cooled pipe. If the ports are built of silica brick, a silicate of iron, calcium, and manganese is gradually formed that has a fusion temperature of about 1100° C., or 400° to 500° below the fusion temperature of good silica bricks. Because of this action, water pipes embedded in the brickwork may be left quite bare, the surrounding bricks being fluxed and then melted away.

No device will render unnecessary careful furnace practice, whether it be a matter of design, refractory materials, water-cooled appliances, or other feature. To really determine the value of a new thing in an open-hearth furnace, as in many other things, the design, materials and furnace practice must be reduced to a standard and then the new feature may be incorporated and its results ascertained. This procedure is perhaps never followed. New things are tried either by themselves or a number at a time accompanied, it may be, by changes in running methods

* Received Feb. 20, 1919.

so that the effect of any one change is obscured. Consequently, various opinions are arrived at regarding the different elements of the case and as a result furnaces are varied to meet conclusions reached so that no two are alike. Experience is gained so slowly because of the length of time required to make a running test that such a state may be expected to continue far into the future. The cooling influence of a water-cooled ½-in. (12.7-mm.) steel-plate surface in a steel furnace extends perhaps 4 in. (10 cm.) in silica brick and will protect the unaltered brick to some extent up to that distance. Brick, however, that has received accretions of oxides of iron and calcium and been fluxed thereby, may be melted or wasted away almost in contact with a water-cooled plate such as the wall of a pipe, as already stated.

A favoring circumstance in the use of a large water-cooled surface, such as the port cooler described in the paper, is that a layer of material from the bath accumulates on it which if only ½ or ¾ in. thick, as it may be, lessens materially the cooling effect of the device on the charge.

Water Supply.—An important requirement for water-cooled devices is an ample supply of clear water without too much lime, acid, or salt in solution or solid impurities, meaning sediment or rubbish, such as leaves, grass, or twigs. Acid water, like that of the Monongahela River which receives the drainage of many coal mines, sometimes corrodes through wrought pipes or plates ⅜ in. (9.5 mm.) thick in a year. Sea water containing about 2000 gr. per gal., or about 3 per cent., of salt corrodes both wrought and cast iron more rapidly than does fresh water. Brackish water is used in cooling both for blast and steel furnaces with no undue troubles. The salinity of the water varies in one case from 200 to 250 gr. per gal., or 10 to 12 per cent. of that of average sea water; the only special precaution observed is to maintain a sufficiently strong flow to prevent any material rise in temperature. As stated in the paper, when the cooling water contains much lime, the latter may be precipitated, upon heating to a certain temperature, and form scale resembling boiler scale within the device. To prevent this the precipitation temperature should be determined and the flow of water so regulated that the temperature of the outflow is well below it. In one case the lime precipitated at 143° F. (62° C.) so the outflow was kept below 130° F. (54° C.).

Sediment, or mud, in water is liable to be deposited at any point in the device where the circulation is slow, in which case the metal will become heated, break, and leak. This quickly happens when the place where the sediment is deposited is exposed to the full heat of the furnace.

Rubbish is liable to check or stop the flow of water, so the furnacemen should watch the outflow for any diminution of volume without change in the valves, as such diminution would indicate a stoppage somewhere and call for prompt cleaning. At one large works, no globe water valves for this purpose are used for they are more likely to be clogged by rubbish than plug cocks, because the water flows straight through the latter.

The possibility of regulating by the reversing valves, described in the paper, the proportion of the outgoing gases that pass through each regenerator appeals strongly to the present writer, as the desirability of having this under control appeared in 1880, when the gas regenerators of the furnaces then in use were always much hotter than the air regenerators and were consequently damaged more in running the furnace.

ROBERT M. KEENEY,* Portsmouth, Ohio (written discussion†).—Mr. Coffin has written a most interesting paper that covers most of the uses to which water-cooled devices are subjected; but to meet with the operating department's approval, these devices must effect a saving in cost per ton of ingots produced in the furnace.

The making of steel in the open-hearth furnace is dependent entirely on the personal element. It is not possible to charge the furnace with so much raw material, turn on a valve for heat, and by use of recording thermometers, clocks, etc. empty the furnace when a steel is made. Competent men trained to the work must be in charge all the time, and the quality and quantity of steel produced is quite as dependent on the care and working qualities of the furnaces as on the raw material and working of the charge itself. Steel making is hot, hard work and the inducement offered to technical men to take up this work is not sufficient to induce many to enter the field. The result is that most of the workmen in an open-hearth plant, the men who actually operate the furnace and make the steel, are not technical men and do not understand the "whys" and "wherefores" and care less. All this means that a furnace must be built as nearly fool-proof as possible. By this I mean that it is necessary to protect those parts of a furnace most subject to abuse by water-cooled devices whenever possible, and these water-cooled appliances must be as nearly fool-proof as they can be made.

The control of the gas entering a furnace is of the utmost importance. This means a properly designed port must have the same design throughout the whole run of the furnace. This is not possible with the ordinary silica-brick port, as the outgoing gases will melt away the brick at the nose so that soon the furnace will have a port very different in design from the original, which will allow the flame to burn the roof, side walls, and, many times, the gas checkers. By using a water-cooled port, there will be a proper distribution of the gas all through the furnace run.

There is an old saying, "You cannot get something for nothing," so the question arises as to what will be the cost of this furnace? Our experience has been that a properly designed water-cooled port does not increase fuel consumption. It will not make a furnace work faster for one heat; in fact, it will probably make the first 50 heats as fast with dry

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†Received Apr. 12, 1919.

ports as with water-cooled ports, but the furnace with water-cooled ports will make from 200 to 300 heats in much less time than a dry-port furnace and with a very considerable saving in fuel and refractories. When it is necessary to rebuild the furnace, the water-cooled port is still there, which effects a saving in rebuilding cost.

The bulkheads of an open-hearth furnace are another cause of trouble. If not water-cooled they burn out quickly and the silica runs into the slag pockets, mixing with the oxides depositing from the waste gases. This causes the slag pockets to fill rapidly with a very hard granite-like material. In rebuilding the furnace, very often digging out this slag is the longest job and keeps a furnace idle longer than would otherwise be the case. Water-cooled bulkheads will decrease the deposit immensely and at the same time effect a saving in labor, material, and lost time during the furnace run.

Another source of trouble in an open-hearth furnace is the burning away of the division wall between the air and gas uptakes. This is particularly true of a plant that has enlarged its furnaces and has not room to bring the gas uptakes up behind the air uptakes but finds it necessary to bring up these uptakes side by side. A cooler placed in these division walls will keep the gas and air from mixing, thus preventing premature combustion, and the cooling effect on the gases is negligible.

The operating man must exercise judgment in applying these water-cooled devices so that the waste water from one cooler may be used for some other cooler and thus reduce the amount of water used. At the same time care must be taken that the piping is so arranged that the clogging or breaking of the pipe will not cause damage to the equipment.

Just where water-cooled devices are a saving, and where they begin to cost more than the returns justify, is a question each plant management must determine for itself. Continuous operation at a good tonnage rate per hour, with a minimum amount of delays for rebuilding, means high production, low fuel, low rebuilding, and low labor costs. It is a a hard matter to determine just what the comparative costs are as between a furnace partly water cooled and one with all possible appliances used because of the difference in charging, kind of steel made, fuel used, and unavoidable delays. However, from experience in our plant here, we know furnaces with water-cooled doors, frames, ports, bulkheads, and skewbacks show: Less average time to make heat during a complete run, less coal per ton of steel, lower refractory cost, lower rebuilding cost, and lower cost of labor and material in repairs. We have taken into consideration the cost of the original installation and the cost of the cooling water in the above statement.

If any one expects a cure-all for every furnace trouble by simply adding water-cooled appliances, he will be disappointed. These ap-

pliances must be designed intelligently, put in intelligently, watched, and taken care of. They should not be exposed directly to the heat but protected with brick. They must be properly cleaned and flushed. Incidentally, it has been our experience that usually twice the necessary amount of water is used by them because it is easier for the workmen. To use excess water means a high pumping cost and a greater heat loss if the coolers are exposed.

As stated in the beginning of this discussion, steel making is hot, hard work. Anything that can be done to alleviate these conditions will return interest on money invested to the operators. Common sense tells us that a furnaceman who does not have to do as much work on ports, as much patching of blocks, front walls, and back walls, while his heat is melting down, will be in much better condition to give the actual making of the steel closer attention, which results in a better quality of product. Water-cooled appliances help to give us these conditions.

Our experience tells us that to determine what water-cooled appliances really effect a saving it is necessary first to equip the weakest part of the furnace with suitable coolers. It is advisable to equip three or four furnaces to get a true comparative cost, as a single furnace might be affected by unusual operating conditions that would give misleading results. The next run of the furnace will show whether a saving has been made and also what is the next weakest part of the furnace. This part can then be equipped and the process repeated until a furnace is obtained that will give maximum continuous service.

Effervescing Steel

BY HENRY D. HIBBARD, S. B., PLAINFIELD, N. J.

(Chicago Meeting, September, 1919)

FOR the purpose of this paper all steels will be divided into two divisions: effervescing and non-effervescing. This classification must be borne in mind as many statements true of one class are not true of the other.

Effervescing steels are those that effervesce in the molds, evolving large volumes of gases, which, in escaping, throw up a spray of molten particles of steel and give the molten metal in the mold a rapid stirring or churning motion until it freezes, or, at least, becomes pasty. This effervescing is intended and proper. The non-effervescing steels, which are more or less completely killed, should not evolve any gas in the molds. Many years ago the writer proposed to call the two classes of steels evolution and solution steels, depending on whether the gases in the molds are evolved or kept in solution, but the present designations seem to be better.

In America, at least, effervescing steels are in a class by themselves and include steels made for boiler, tank, and ship plates, steel pipe, soft steel wire, wire nails, soft machinery, and structural steels. These are all low-carbon steels with less than 0.40 per cent. carbon. Low-carbon steels that do not effervesce but are killed are also made; these are put into ingots and castings but they are outside the limits of this paper. Effervescing steel is cast into square or slab molds, either top or bottom poured.

Most of the effervescing steel is made in the basic open-hearth furnace, though much is made by the acid Bessemer process and a little is made in the acid open-hearth. It is not made by the crucible or electric-furnace process, in which the volume of gas required cannot be produced in the steel and evolved from it in the mold to give the effervescence desired.

Any steel ingot, when teeming is finished, is of a certain size or volume. Leaving out of consideration for the moment the lessening of the over-all dimensions of the ingot through contraction due to loss of heat, the ingot may be placed in one of three classes: It may settle, losing in volume; it may stand, that is, remain constant in volume; it may rise, that is, increase in volume.

Into which class an ingot will fall depends on the gases of the steel,

their escape from the metal during solidification, and the holes they form in the ingot. These three classes have no strict lines of demarcation but can be distinguished from one another sufficiently well for practical purposes.

Gas Holes

The gas holes in effervescing steels are of three kinds, each of which occurs in a distinctive manner. Skin holes are at or near the outer surface. Intermediate holes, sometimes called deep-seated holes, are in a zone between the surface and the center. Central holes are located at random in the central portions.

Skin holes constitute defects and are not desired. Intermediate and central holes exist in good standing and settling effervescing steel but are of such size and are so located as to give no trouble; neither do they lower the quality. The gases in the three kinds of holes have not been separately analyzed. There is some ground, however, for believing that the skin holes are formed chiefly by hydrogen, the intermediate by carbon monoxide, CO, and the central by nitrogen and ammonia, NH_3.

The correct behavior of good effervescing steel in the mold is this: As soon as the first of the steel enters the mold, whether top or bottom cast, it evolves a myriad of gas bubbles, the evolution continuing until the ingot has become solid. Each bubble, as it reaches the surface and escapes, is followed by small drops of melted steel thrown up perhaps 2 or 3 ft. (0.6 or 0.9 m.) by the inrushing metal as it fills the evacuated space. These drops burn more or less completely in the air and collectively make a brilliant, if small, pyrotechnic display. This effervescence reaches a maximum when teeming ceases and then gradually lessens until the ingot is covered or, may be, it is all solid. The steel begins to freeze in contact with the mold almost at once after teeming is finished and the freezing walls progressively grow in thickness by accretion from the still liquid and effervescing central metal. This process of solidifying is called in the shop "rimming in;" and when everything is right, the rim has a level top. When the rim is 1 or 2 in. (2.5 or 5 cm.) thick a cast-iron cover an inch or more thick is laid to prevent a rising of the central metal, which would usually take place (if it were left to itself) just before final freezing, and the ingot has a flat top. In such an ingot there will be practically no skin holes. The ingot will consist of clean solid steel to a depth of 2 or 3 in., depending on its size and carbon percentage, then will come the zone of intermediate holes, and within them the central part containing a few scattered gas holes; see *B*, Fig. 1.

The proneness of steel to effervesce in the molds is in a way a function of its carbon content. The lower the carbon the more its rimming-in capability, and when the carbon is extremely low, say 0.04 per cent. or less,

the steel does not pass through a mushy stage before final freezing and the ingot may rim in to the center and not require to be capped as is customary with most effervescing steels.

Effects of Too Mild an Effervescence.—When the effervescence is too mild, the steel rises in the mold after teeming has ceased. This shows that skin holes are forming, which displace their bulk of metal and cause the rising. In pronounced cases, the steel continues to rise, when teeming is finished, for several minutes. The holes are covered by only a thin wall of steel, or, as they say in the shop, the ingot is "thin skinned." The holes may be 2 in. (5 cm.) or even more in length and as close together as the cells of a bee's honeycomb, see Fig. 6. Steel containing such holes must be scrapped as it cannot be used for any purpose where any regard is given to the surface. A plate or bloom rolled therefrom is covered with defects such as tears, pits, seams, and scabs.

Such a profusion of skin holes results from a much too high casting temperature and insufficient oxygen in the metal, which latter is evidenced by too gentle boiling of the bath. In the mold there is little or no effervescence and the steel has an oily appearance. The displaced metal that rises may amount to 12 or 15 per cent. of the volume of the ingot and perhaps overflow the mold. If the ingot is bottom cast, the spot where the sprue is broken from the ingot will show long skin holes radiating from the center to the circumference of the fracture.

In less pronounced cases, the skin holes may be numerous and extend somewhat irregularly over the whole ingot surface but deeper seated with thicker skins over them than in the worse cases. Then, if they are not burnt into in the heating operation, the rolled product may have a good surface; or a part may be burnt into by irregular heating, which will cause pits in the plate or seams in the bloom in that part; see *C*, Fig. 1. In other mild cases, a few small skin holes may occur in a small fraction of the ingot surface, but the pits they cause will be enough to condemn any plate containing them; see *A*, Fig. 1.

Steel that Pipes or Settles After Teeming.—Piping or settling of effervescing steel after teeming is caused by the escape of such volumes of gases that a greater bulk of the liquid is displaced than is represented by the residual gas holes. Steel has been made by the writer that settled to about one-half the volume it had at the end of teeming leaving a shell, pipe, or "bootleg," that had solidified against the ingot wall. That was an extreme case in the early days of the basic open-hearth process, when the effect of free oxide of iron in the slag was not foreseen from the results of acid practice. Such an ingot will have intermediate and central holes below the shell or pipe and may behave well when rolled but it will weigh less than was intended and will give an unduly large percentage of scrap.

Genesis of Gas Holes.—The actual genesis of the gas holes in effervesc-

ing steel is a matter for speculation but, assuming the gases forming them to be as previously stated, the following suppositions may seem reasonable as some of the facts support them. Of course, the gases that escape from the metal in the mold, or at least their ingredients, are contained in the steel when it enters the mold. The atmosphere takes little part for it is immediately expelled from the mold by the gases evolved from the first metal to enter after which only the stream from the ladle is exposed to it, and that for a small fraction of a second.

Carbon monoxide, whether in solution as such in the liquid metal or formed just before it escapes, as has been claimed, plays the most important part. In a normal case, this gas begins to be evolved as soon as the first steel enters the mold, because apparently the solvent power of the metal for it is lessened by the lowering of its temperature. This escape of carbon monoxide continues until all the metal of the ingot has congealed, when the last of it is expelled.

Müller found that carbon monoxide is insoluble in cold steel. As the ingot walls thicken, the liquid central metal (with carbon over 0.10 per cent.) becomes somewhat mushy and the evolution of gas therefrom gradually diminishes. The bubbles forming the intermediate holes are then made, presumably of the last carbon monoxide to be liberated, and lodged along the face of the solid wall entrapped by lack of fluidity of the mushy metal. It might be thought that these intermediate bubbles, as well as the skin holes, are formed by hydrogen, which looks reasonable in some cases where they are nearer the surface than usual, were it not for the fact that sometimes both skin holes and intermediate holes exist in the same ingot distinctly apart, which is fair evidence that they are formed by different gases, see *C*, Fig. 1 and Fig. 3. The intermediate holes have clean bright surfaces and, ordinarily, when not unduly large are closed and welded by rolling.

Carbon monoxide, in escaping from the steel in the mold, very likely carries off some of the hydrogen, nitrogen, and ammonia, but the small residues of these gases are sufficient to form at least a few holes when conditions are favorable therefor. The reason why no carbon monoxide is evolved in the mold when the metal is too hot is not apparent.

Of the hydrogen dissolved in the metal, a relatively minute quantity, compared with the enormous volume of carbon monoxide in normal steel, seems to remain in the metal until the instant of solidification. It is then set free and forms bubbles that cling to the ingot wall; these are small at first but, if not dislodged, grow on the inside as the ingot wall thickens by progressive freezing. Each bubble thus acquires an elongated shape with its axis normal to the ingot wall. When, however, the evolution of carbon monoxide is sufficiently brisk, these hydrogen bubbles are washed off mechanically as they form by the rising carbon monoxide bubbles, which they join and rise with them to the surface and

escape. All or only a part of them may be so eliminated, according as the casting conditions are correct or not but any that remain form skin holes. This explanation is supported by the fact that skin holes are found sometimes at or near the bottom of the ingot, where the washing action of the carbon monoxide is weakest, when there are none higher up in the ingot; see Fig. 2. In extreme cases skin holes continue to grow until the unfrozen metal in the central portion of the ingot has set free all its hydrogen and has become pasty.

Residual nitrogen and ammonia may be present in even smaller quantities than hydrogen and be progressively concentrated in the last metal to remain liquid at the center of the ingot, separating then to form the central holes. These always have bright surfaces and are not to be readily detected in the rolled product being, as a rule, probably welded in rolling, the gases being reabsorbed by the metal under the great pressure of the rolls.

Between the ordinary, or normal, and the extreme cases cited any gradation may occur. If the steel is slightly too hot or not sufficiently active in the furnace, effervescence may apparently be as brisk as in the normal case but the metal rises slowly as it "rims in" forming a frustum top having a volume of 2 or 3 per cent. of that of the ingot. In this case the skin holes will be small and few in number and may have a fairly thick skin of sound metal outside, say, ½ or ¾ in. (12.7 or 19 mm.) thick, enough to hold during heating and rolling so that the plate or bloom has a clean surface.

If the temperature is still higher or the boil is still less active or if some gas solvent is present, the effervescence at first may be retarded somewhat and then become more brisk. The steel rises in the mold faster at first than "rimming in" proceeds so that that operation only begins to show after a minute or two or perhaps even longer, when the rising is slackened. A frustum top is then formed, as in the case just cited, and the total rise may equal 5 or even 10 per cent. of the ingot volume. A tendency to skin over may be noticed as soon as teeming has ceased, floating islands forming and then remelting more or less completely. Steel that behaves in this manner is likely to have surface defects in the rolled product; and pretty sure to if the ingots are allowed to become cold and then reheated for rolling. The time this takes in the heating furnace is longer than when the ingots are put in hot from the casting heat, causing a deeper oxidation of the surface metal, which is likely to uncover the skin holes present, oxidize their interiors, and cause surface defects in the product. The skin holes will be larger and more numerous than in the previously cited case as well as being nearer the surface, sometimes within ¼ or ⅜ in. (6.35 or 9.52 mm.).

Effect of Stirring Steel Bath.—According to these explanations the carbon monoxide plays an important and beneficial part in making good

effervescing steel. An effort has been made to "shake out" this gas before casting by stirring with many cold steel bars or rods, which will cause the outrush of great volumes, but such practice has been abandoned or cut down to the use of a few, not over five or six rods. At one works, many years ago, it was customary to use thirty rods to a 20-ton heat to drive out the gases that, it was assumed, injured the quality of the steel; but the results were not any better and the practice was discontinued. At that works the custom was, as soon as teeming was completed, to fill a mold with sand over which a steel plate was clamped down by a wedge driven under a bar passed through the lugs. Therefore it was not known if the steel would have effervesced and rimmed in properly or not, but it is a fair presumption that the steel was not properly made or the haste with which the molds were filled with sand would have been found unnecessary. The object of such haste was to prevent the steel rising in the mold, which the gas in forming bubbles in the steel tended to cause. This evolution of gas proceeds with considerable force so that jets of liquid steel were sometimes forced out through the sand endangering the workmen; at other times the molds were lifted from their stools and even toppled over, which required a pressure of at least 50 lb. per sq. inch.

The active boil in the furnace necessary to give good effervescence in the molds is the escape of carbon monoxide, so it would seem that unlimited stirring should not be given the bath lest the action in the molds may lack the necessary vigor and skin holes in the ingots result. The carbon monoxide, no doubt, carries off other dissolved gases when it escapes in the furnace so there will presumably be more hydrogen, nitrogen, and ammonia present in the unfinished steel when the boil is gentle; consequently, the holes they form are liable to be larger and more numerous. Thus a normal or proper boil in the furnace gives in the mold little hydrogen, which is easily eliminated, while a slack or weak boil gives too much hydrogen, which forms the profusion of skin holes with thin skins.

All three kinds of gas holes may exist in one ingot, but usually only when the skin holes are comparatively small and few in number, see Fig. 3. An ingot with a pronounced case of skin holes may be quite sound within, see *A*, Fig. 1. Whether or not the central metal has good physical properties cannot be told as the ingot is always remelted. The stirring due to the effervescence aids apparently the agglomeration of the sonims (sulfides and silicates) in the metal, which then rise to the surface leaving the steel so much the cleaner and thereby improved in quality. It has been considered that good thick-skinned steel without skin holes has better physical properties, particularly ductility, with the same amount of hot-working than steel containing them or killed steel of the same composition both of which will retain a substantial part or all of the sonims in the steel when it entered the mold. Proof of this is

lacking and is perhaps only to be had from the averages of a large number of tests of all three kinds of steel.

When larger ingots are made, requiring an additional heating and rolling operation, the extra hot deformation resulting tends to improve the ductility. Such larger ingots are more liable to have surface cracks than smaller ones but the extra hot-working tends to obliterate these by scaling and rolling so that the product, either bars or plates, is merchantable. Segregation is generally more marked in the larger ingots, which fact must always be considered.

The presence of sonims probably increases the susceptibility of steel to corrosion as each non-metallic particle forms an electro-negative spot, which becomes a center of corrosion when conditions favor that operation.

Whether a tear or crack in a bloom of low-carbon steel is due to red-shortness or to skin holes may usually be told by examination of the walls of the break, at least in blooms 6 in. square or more. If the steel is red-short the sides of the flaw will show granular surfaces while if skin holes have caused the defect they will not be wholly obliterated but may be seen within even though distorted and drawn out by the elongation of the piece in the hot-working operation.

Testing Soft Steel for Homogeneity.—To show the lack of homogeneity of soft steel, a flat bar or strip may be bent 180°, shut down, and then the bend opened until the piece breaks. The fracture shows something of the structure at that point and a plate may show flattened seams that have opened and are the remains of gas holes in the ingot. They may or may not be welded. Another way is to soak the bar or strip for several days, or a week, in a beaker of hydrochloric acid made by diluting, say, 250 c.c. of commercial acid with an equal bulk of water. Each piece should be treated by itself in a separate vessel to prevent electrolytic action between different pieces. The acid will penetrate deeply where the metal is electropositive and show the lack of uniformity the ingot possessed.

Making Effervescing Steel in the Basic Open Hearth

Running the furnace in making effervescing steel is like ordinary practice except that the kitchen, or laboratory, must be hotter at the end to provide for the higher fusion point of low-carbon steel. One per cent. of carbon lowers the fusion point of iron about 90° C. A bath with 0.10 per cent. of carbon should be about 72° C. hotter than if it contained 0.90 per cent. To ensure this the kitchen must be still hotter, perhaps 90°, than when 0.90 per cent. carbon steel is being made.

When melting a cold charge, the full power of the furnace should be employed. When all is melted the temperature should be raised gradually to the proper degree for tapping at the end. The carbon content, when melted, is preferably about 0.50 per cent. Ore is added from time

to time to oxidize the excess of carbon until the content of that element is 10 or 15 points above that desired in the finished steel, which excess is oxidized and boiled out by the oxide of iron present in the slag. The final boil desired is not easily described; but with 0.20 per cent. carbon, the whole surface of the slag is in motion from the bursting bubbles of carbon monoxide. For steel to be killed, such action would be too vigorous. The briskness of the boil lessens as the carbon in the metal diminishes; and when it is very low, the boil consists in relatively few scattered bubbles over the surface.

The casting temperature is the chief concern of the furnaceman. The proper degree is that which permits the metal to be teemed cleanly into the molds with the formation of an incipient skull in the ladle not large enough to cover the ladle bottom. Appearances in the kitchen and taking tests give the knowledge for its control.

The proper effervescence in the molds demands the proper boil in the furnace. If it is too gentle, more ore must be added; if too brisk, more time with perhaps the addition of pig iron or manganese alloy, preferably spiegel, is required. The addition of aluminum in the ladle will check a too brisk action and, in some works, it is added regularly up to 5 oz. to the ton. If the steel shows a settling tendency in the molds, the excessive effervescence that causes it may be checked by throwing in a little aluminum, say, 1 or 2 oz. to the ton.

Making Effervescing Steel in the Acid Open Hearth

In the acid open hearth, the boil is, as a rule, not so vigorous as in basic steel of the same carbon content; and while sufficient effervescing action can be had, it rarely happens that such action is excessive. Hence the precautions and treatment to prevent settling or piping followed in basic steel are rarely or never needed in acid steel. No aluminum or silicon are needed or used.

To get a proper boil there must be in the slag enough unspent oxide of iron to give it an earthy fracture and black color when poured in a cake $\frac{1}{4}$ or $\frac{5}{16}$ in. (7.9 mm.) thick on an iron floor, while that which chills on the handle of the test cup, about $\frac{1}{16}$ in. (1.5 mm.) thick, will be black and vitreous. With a green or yellow vitreous slag, the steel is likely to rise in the mold and have skin holes because of too gentle effervescence.

Method of Casting Effervescing Steel

While effervescing steel may be either top or bottom cast, that does not mean that it is a matter of indifference how any given steel is cast. A melt may behave well when bottom cast that would rise in the molds if top cast. The chief difference in the two methods is in the rate of filling the mold. A group of bottom-cast ingots weighs perhaps 15 to 20 tons

so that the slow rate of filling allows the metal to cool and in so doing to evolve the carbon monoxide that causes the effervescence. If the steel is to be top cast, it must be more strongly oxidized in the furnace and so have a more vigorous boil and be cast slower either by pouring two or more molds at once, so as to fill them more slowly, or by the use of a nozzle of small diameter relative to the cross-section of the ingot or ingot group if the steel is bottom cast. A 2-in. diameter nozzle is right for a 10- to 15-ton ingot or group.

Plate steel that is to be rolled into a finished product at one operation is bottom cast because that favors the production of thick-skinned ingots and lessens wrinkles and cold shuts in the ingots, which might make defects on the surface of the plate.

If the steel stands and rims in properly, the covers may be put on when the rim is 1 in. (25.4 mm.) thick, which requires 2 or 3 min. After ½ hr. the ingots may be stripped.

Bessemer Effervescing Steel

Soft effervescing steel for pipe, wire, and other purposes is made by the Bessemer process, its low carbon content particularly favoring its use for purposes when it is to be welded, as in pipe. The two important conditions of manufacture of such steel are that it shall be blown "full" and "close." "Full" means that it be more fully blown than steel for other purposes for which the Bessemer process is used, blowing continuing 2 or 3 sec. after the last luminous streaks have disappeared from the flame. This slight overblow gives the metal the proper proportion of oxygen, which makes and liberates the necessary quantity of gas to give good effervescence. If turned down "younger," the steel is likely to rise in the molds. "Close" means that the casting temperature is near the freezing temperature, as in the case of open-hearth steel mentioned previously. This calls for close control of the temperature by accurate scrapping and blowing. A tendency for the steel to effervesce excessively so as to settle in the mold may be corrected by the addition of a little gas solvent, say 6 oz. of silicon or ½ oz. of aluminum to the ton for mild cases and more for more pronounced ones. If the steel settles moderately it will probably roll all right without developing surface defects, but flat-topped ingots are preferable for they will work up into finished forms with less scrap. When soft Bessemer steel, like pipe steel, fails to effervesce properly in the molds, a few small pieces of limestone thrown in cause a sputtering and keep the top "open," favoring the evolution of gases from the metal.

It has been claimed that the presence of sulfur, say 0.06 per cent., aids greatly in making soft Bessemer steel effervesce properly in the molds. At one works where direct metal with 0.02 per cent. sulfur was blown,

it was difficult to get a good action in the molds, but when the same metal was remelted in a cupola and had thereby been given more sulfur, the desired effervescence was obtained.

Plate Steel

A steel plate must have, in addition to the desired physical properties, a clean surface free from visible defects of all kinds. There are two ways

Fig. 1.—Fractures of soft plate steel ingots made by the author at the Linden Steel Works in 1888.

of rolling plates: the ingot may be of such size and shape that it may be rolled direct from the ingot to the finished plate at one operation, or the ingot may be larger and so must be rolled into a slab, which is cut up into pieces which are reheated and rolled into finished plates on another mill.

For direct rolling the structure and surface of the ingot must be right. It must contain no skin holes and the intermediate holes should be, at

least, 1 inch, and preferably 2 inches, from the surface. The ingot should have a flat top and no surface cracks.

For double rolling, the soundness of the ingot or the perfection of its surface is not so important, as small defects, such as shallow pits, snaky cracks, and cold shuts, will usually be obliterated by the two heatings and rollings. So the steel need not be so carefully made for double as for direct rolling. The double rolling tends also to improve the physical properties and so make up for some lack of quality. The larger ingots made for double rolling are liable to have more pronounced segregation and irregularity of composition than smaller ingots.

FIG. 2.

Figs. 1 to 6 show split and broken ingots and slabs illustrating some of the internal features of plate ingots. Fig. 1 shows the six broken fractures of three bottom-cast slab ingots, 8 by 20 in. (20 by 50 cm.) in cross-section made by the acid open-hearth process. The steel had about 0.12 per cent. carbon, 0.35 per cent. manganese, 0.05 per cent. sulfur, 0.06 per cent. phosphorus, and 0.01 per cent. silicon and was intended to effervesce in the mold. The ingots were broken in two under a drop. The top ingot *A* had a few skin holes near the corners indicated by the arrow-heads, which caused pits in plates rolled from the other ingots of the same heat. The remainder of the ingot was practically solid and the ingot probably weighed more per unit of height than the other two, which were thicker-skinned, such holes as they contained being located deeper in from the surface.

The next ingot *B* was about ideal, there being no skin holes and the intermediate holes having an outer covering of over 2 in. (5 cm.) of clean solid steel. Such an ingot may be rolled into a practically perfect plate.

The bottom ingot *C* had a zone of skin holes with ½ in. (12.7 mm.) skin of solid steel outside. Within the skin holes is a zone of intermediate holes. This steel was cast slightly too hot and rose a little in the mold but not ruinously so. Unless the time in the heating furnace is prolonged unduly, which might cause the outer skin to be burned through, such an ingot may be rolled into a marketable plate.

Fig. 2 is a vertical section through the center of a good slab ingot

18 in. (45.7 cm.) wide. This ingot has a flat top and the usual zone of intermediate holes, while across the bottom and a little way up on each side is a partial zone of skin-holes with ½ in. (12.7 mm.) of sound steel outside. This ingot furnishes evidence of the explanation previously given as to the mechanical action of the carbon-monoxide bubbles in washing off the forming hydrogen bubbles. Such action is manifestly weaker on and near the bottom than farther up. The size and number of rising bubbles increases as the top is approached.

FIG. 3.—LARGE OPEN-HEARTH INGOT.

Fig. 3 is the vertical section of an ingot 13 in. (33 cm.) wide of soft, basic, open-hearth steel having 0.08 per cent. of carbon. This ingot has all three kinds of gas holes previously referred to though the skin holes are so deeply located as to impair but little the quality of the steel and its suitability for its intended purpose. This steel had too feeble effervescence in the lower part of the ingot to dislodge the hydrogen gas bubbles forming the skin holes.

Fig. 4 shows the broken section of a slab 4 in. (10 cm.) thick, which was rolled from an ingot 13 by 22 in. (33 by 55.8 cm.) in a two-high mill running one way, the ingot being passed back over the top. This was an excellent ingot without skin holes, the zone of intermediate holes being 2 in. in from the surface and the bubbles small and relatively few in number.

Fig. 5 shows another slab of the same size and from the same sized ingot as Fig. 4. This ingot was only fair as the skin is but ½ in. (12.7 mm.) thick though the plate rolled from it might be merchantable. The right-hand side of this slab shows a heavy zone of skin holes, such as sometimes will not be completely welded up in the rolling operation. When that happens, they are liable to be cut through in shearing the plate to size and a split edge is disclosed, which is a serious defect. Traces of intermediate holes may be discerned but the holes on the top and bottom sides have been obliterated by rolling.

Fig. 6 is of a typical badly pitted ingot 12 by 42 in. (30 by 106 cm.) in cross-section with skin holes extending 2 in. in from the surface. It contained 0.20 per cent. carbon, 0.036 per cent. sulfur, 0.011 per cent. phosphorus, and 0.40 per cent. manganese. It was cast too hot and was useless, except as scrap.

When a soft ingot having a section or area infested badly with skin holes is stripped red hot, such a section or area will show because it loses heat and becomes black quicker than the remainder of the ingot. Such

a quicker cooling part, having less metal per unit of length, will, when the ingot is rolled to a plate at one operation, cause the plate to be narrower

FIG. 4.—SLAB MADE AND PHOTOGRAPHED AT NORWAY STEEL & IRON CO.'S WORKS. PLATE STEEL ROLLED ONE WAY FROM 13 BY 22 IN. TO 4 BY 22 IN. EXCELLENT SLAB. BLOWHOLES VERY FAR IN. MADE IN 1884.

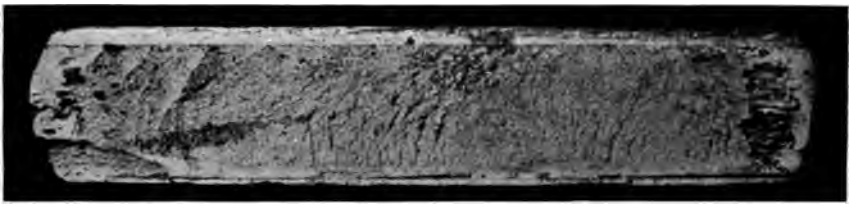

FIG. 5.—SLAB MADE AND PHOTOGRAPHED AT NORWAY STEEL & IRON CO.'S WORKS, SO. BOSTON, MASS. PLATE STEEL (BOILER) ROLLED ONE WAY ONLY FROM 13 BY 22 IN. TO 4 BY 22 IN. RATHER POOR STEEL; BLOWHOLES TOO NEAR SURFACE, BUT NO PITS. MADE IN 1884.

there than elsewhere where skin holes are fewer or absent, and the plate will be of irregular shape. A good slab ingot will, when properly rolled,

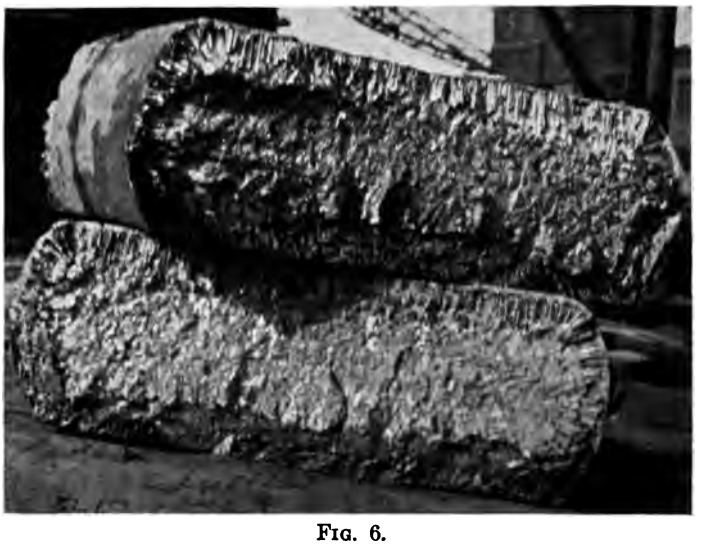

FIG. 6.

give a rectangular plate as wide at the top as at the bottom, showing a substantially even distribution of such gas holes as it contains.

The appearance of the rolled plate gives some indication of the temperature at which the rolling was finished. If at a bright red heat, say about 750° C., the scale will be set and be of the dark slate color of magnetic oxide of iron (Fe_3O_4); plates $\frac{3}{8}$ to $\frac{1}{2}$ in. (9.52 to 12.7 mm.) thick should have such a color. Finishing at a lower temperature, but still visibly red, will cause the scale when cold to have the red color of Fe_2O_3; plates $\frac{1}{4}$ in. (6.35 mm.) thick or less usually have this color. On heavy plates, 1 in. thick or more, which are likely to be finished hotter, say, up to 900° C., the scale may rise and loosen itself more or less completely from the plate after rolling is finished.

Manufacture of Steel Rails

BY ROBERT W. HUNT, CHICAGO, ILL.

(Chicago Meeting, September, 1919)

THE American Institute of Mining and Metallurgical Engineers was the first American technical organization to consider steel-rail specifications and sections. If I am not mistaken, the first contribution on those subjects to the Proceedings of the Institute was presented by Prof. Thomas Egleston, in 1874, and was entitled "Investigations on Iron and Steel Rails made in Europe in the Year 1873.[1]" The next contribution was Dr. C. B. Dudley's report of his investigation for the Pennsylvania Railroad of the wear of steel rails in their tracks, in 1878, which contribution led to a very memorable discussion that extended through several meetings.[2] In 1880, Mr. C. P. Sandberg[3] presented an article on rail specifications and inspection and rail sections. This was followed, in 1881, by A. L. Holley,[4] in his exhaustive report on the then used American rail sections. The next contribution, I believe, was made by myself[5] at the Buffalo meeting, in 1888, on steel-rail specifications, which I followed,[6] in 1889, by a paper on steel-rail sections. Since that time there have been numerous contributions to the *Transactions* of the Institute in relation to the manufacture of steel rails.

The American Society of Civil Engineers officially took up the investigation of the relation of rail sections to the wear of car wheels in 1888. In the prosecution of this work, the Society appointed a special committee, whose labors extended through several years and resulted in an exhaustive report to the Society in 1893.[7] Later, another committee was appointed to consider specifications for the manufacture of steel rails; the labors of this committee extended over a period of several years, a final report[8] being made in 1910. It was my fortune to be the secretary of both of those committees.

The American Society for Testing Materials, organized in 1899, took up the consideration of rail specifications and is still giving attention to that subject through its established procedure of investigations by committees.

In 1900, the American Maintenance of Way Association was organ-

[1] *Trans.* (1874) **3**, 44.
[2] *Trans.* (1878) **7**, 202.
[3] *Trans.* (1880) **9**, 193.
[4] *Trans.* (1881) **9**, 360.
[5] *Trans.* (1888–89) **17**, 226.
[6] *Idem.*, 778.
[7] *Trans.* Am. Soc. Civ. Engrs. (1893) **28**, 425.
[8] *Trans.* Am. Soc. Civ. Engrs. (1910) **70**, 456.

ized, which society has since adopted the title of the American Railway Engineering Association. Early in its history, and continuing to the present time, the questions of specifications for and sections of steel rails have had the consideration of a special committee.

But in view of the fact that this Institute was the first American technical society to give consideration to the subject, it seems proper that at this meeting, which is being held at the point of the largest production of steel rails in America, and practically on the 75th anniversary of the rolling of the first iron rail and the 54th anniversary of the rolling of the first steel rail in America (and the last event in Chicago), the Institute should have presented to it some thoughts bearing upon the still unsettled questions as to what should be the specifications and procedure governing the manufacture of steel rails. I do not propose to submit at this time a detailed specification, but confine myself to certain points which seem to me to be of great importance.

From the earliest consideration of the subject, the value of proper inspection has always been recognized, and some of the most important points of discussion have been as to how that inspection should be conducted. It has been the writer's fortune to have been connected with the manufacture of steel rails during the whole of the existence of that industry in America and therefore he has had, from practical contact, an intimate knowledge of the various developments that have taken place in that industry; and, particularly, as during the last 31 years he has made a specialty of steel-rail inspection, he feels that he may claim to have had the best of opportunities to become familiar with the details of the procedure.

This familiarity led him to comply with the request of Messrs. Julius Kruttschnit, director of Maintenance and Operation, and John D. Isaacs, consulting engineer, of the then so-called Harriman Lines, that he should prepare and install a more detailed system of inspection than had been used, under which there would be a thorough supervision of the progressive steps in the manufacture of rails; beginning at the making of steel down through the various stages of manufacture to the loading of the finished rails upon the cars for shipment. As this plan met with the approval of several others of the principal clients of Robert W. Hunt and Co., notably of the late Mr. W. A. Gardner, President of the Chicago & Northwestern Railroad Company, that firm established, in 1912, what is known as their "Special Inspection." This consists in placing inspectors in all the departments of the steel works, beginning at the open-hearth furnaces or Bessemer converters, at the soaking pits and the blooming mill, the rail mill, the drop-testing machines, the hot beds, drilling and straightening departments, and, finally, the inspection of the finished rails. These men are placed at each of these several departments day and night,

their duties being to observe carefully all the details taking place; and if there should be any deviation from the provisions of the specifications under which the rails were purchased, or deviations fron the accepted practice of the mills, or abnormal conditions of any kind, to make note of the same; and while not at liberty to interfere directly with the workmen, to call the attention of the foreman of the particular department in which the occurrence is taking place to what is happening, and also to pass a written report to the inspector stationed in the next succeeding department, and so on to the finished-rail inspectors. The chief inspector is also required to give a written report of unsatisfactory conditions to the superintendent of the mill, so that there may be an absolute record of what has happened to each heat of steel made on either turn, be it day or night, and as the rails are stamped with their heat numbers, and location in the ingots from which they were rolled, the mill management will possess an individual history of each rail, with the possible identification of the individual workmen who were in control of the manipulation of the rail from beginning to end.

Moreover, a comprehensive report is rendered to the proper official of the railway company for which the rails are manufactured, resulting in the archives of that railroad possessing a detailed history of the making of the rails. Hence, should a rail fail in service, it will be possible to not only have cognizance of all the details of its manufacture, but the makers will be able to determine the workmen who participated in that manufacture. As can be appreciated, this leads to a personal responsibility being attached to every rail. It must be remembered that the workmen at the mills are paid by the ton, hence they are naturally desirous of producing as large an output as possible, and under the ordinary system of inspection, as soon as the rail left the works, all individual connection with it vanished, or, at best, any process of personal connection was extremely difficult and uncertain.

The result of this special system of inspection has been most satisfactory to both the manufacturers and the purchasers, and it receives the approbation of the workmen themselves, as each man knows that, if he properly performs his duties, he is protected against the danger of being held responsible for the acts of a less careful operative, while the mill management has given it, without cost, a thorough system of supervision. That it has proved satisfactory to the railway companies that have been employing it is best evidenced by the following, gathered from the records of the American Railway Engineering Association.

The statistics covering open-hearth steel rails give the failures that have occurred in 5 years, ending Oct. 31, 1917, on 37,862 mi. of track. Forty-five per cent. of this mileage comprised rails rolled under special inspection, and therefore 55 per cent. was made under other conditions. The total number of failures per hundred track miles on all of the mileage

reported was 31.1, while the number of failures per hundred track miles on the rails covered by special inspection was 26.6, and the number of failures per hundred track miles of those not covered by special inspection was 34.8. Thus, there was 30 per cent. in favor of specially inspected rails.

As each ingot is a separate and individual casting, I believe that, in addition to the tests representing the whole heat of metal, there should be some way of determining the physical character of the steel rolled from each and every ingot. As a destructive test seems to be the only way to actually ascertain that point, and as, of course, you cannot destroy all the rails, I, in 1915, installed what is known as the nick-and-break test, under which a test piece is cut from the top end of the first, or *A*, rail rolled from each ingot. This is nicked and broken, thus enabling the inspector to examine the fractures and, in case of the presence of segregation, pipe, or some other mechanical defect, to reject the rail represented by the test piece. Following such rejection, a test piece shall be cut from the bottom end of the same rail and broken; if that piece has similar defects, the second, or *B*, rail is rejected, and a piece cut from its lower end, and tested. In case of failure, the procedure is continued for the succeeding rails until a sound one has been reached. This has proved an expeditious and cheap procedure, and in my judgment very satisfactory.

A large tonnage of rails made under it is in American railway tracks and a much larger tonnage in those of Canadian railways. In fact, at the present time every standard steel rail rolled in Canada, and for all of that country's roads, is so made.

As the chemical composition of the steel is a fundamental part of rail specifications, a very important point is the taking, for analysis, of the drillings representing the heats of steel from which the rails are rolled. The drillings must be taken from test ingots, and the time and manner of casting these ingots, as well as their shape, is receiving attention. In fact, the whole matter of ladle-test ingots is now receiving the attention and study of a committee of the American Society for Testing Materials in conjunction with the U. S. Bureau of Standards. Pending a full report on the subject by them, it would appear desirable to require the adoption of a standard sized and shaped ladle-test ingot and to insist that it be cast as sound as possible by the addition of aluminum to the molten steel; further, that two or three should be cast at equal intervals in a heat during the pouring of the large ingots and drillings from them carefully mixed. Thus, greater protection would be afforded as against a single small test ingot not representing, thoroughly, say, a 100-ton heat of steel.

All specifications should demand that the inspector representing the purchaser should have the right to witness the taking and mixing of the drillings, from which mixture, upon his request, he be given a portion for

the purpose of check analysis. As a rule this is done, but at some plants it is a subject of dispute, which should be prevented.

A few years ago there were many rail failures through what were designated as crescent-shaped breakages; that is, ruptures in the rail flanges extending from the outside inward toward the web, the broken pieces being of crescent shape. These fractures were frequently followed by complete cross-fractures of the rails, resulting in many costly and oftentimes fatal accidents. It was found that, as an almost inevitable rule, there was a longitudinal seam in the bottom of the rail flange at the point of fracture, which was directly under the web of the rail. This situation caused great anxiety and led to much discussion as to the causes of the seams or laps. One steel company developed and established a deseaming adjunct to its rail mill, by which the outer steel of that part of the blooms subsequently forming the heads and flanges of the rails was milled off. The works claim success from the scheme. The later adopted sections have thicker flanges and the crescent-shaped breaking trouble seemed to have become minimized; but I regret to say that, during the past year, one of the prominent railway systems has been having a large number of such failures from heavy rails. Another system encountered the same kind of defects, which, in this case, were fortunately discovered before serious damage occurred. The rails were from different steel companies; therefore, it is again a matter demanding prompt and effective action. To require all producers to establish de-seaming mills is very radical, but if that is the only way property and life can be protected from such danger, it may have to be done.

Another point of less importance, but still of prime necessity, is the milling of both ends of the rails. This has been the practice of English mills for years and some American makers mill one end of some of the rails made by them. Depending on chipping and filing for the finishing of the ends of the rails is very unsatisfactory. Milling would not only eliminate all fins resulting from the hot sawing, but positively assure squareness of ends and accuracy of lengths, all three of which are of great importance in relation to joints, a part of track maintenance that is receiving more and more attention.

Much thought and considerable experiment have been given to the betterment of the cold straightening of rails. From the time of that procedure being represented by the blows of a heavy sledge swung by a sturdy man, the practice has been a brutal one. To accomplish the semblance of the desired result, the steel has to be bent beyond its elastic limit, thus establishing strains with an ever present danger of causing actual or incipient ruptures in the metal, which may result in complete rail failures. In my judgment, it is practical to do away with cold straightening, by proper hot straightening of the rails. If, in case of uncontrollable conditions, a few rails should come from the hot beds

unsuitable for use, they could, after some cold straightening, be classed as seconds and used for other than main-track purposes.

In the present practice, a rail is put under the cold press and most carefully, but brutally, punched into seeming straight line and surface. Frequently, it is later put on the ties and with equal care spiked out of a straight line into curves of varying degrees. Rails can be hot straightened without any short kinks and with lines that will permit good loading for transportation and satisfactory track laying; thus can the expense and danger of the cold straightening be avoided. There are now at least a thousand tons of rails, finished as above advocated, giving satisfactory service in the lines of several railroads having heavy traffic, and they have had over 3 years of trial. The rail makers will welcome such a departure from present practice and, I am confident, make it successful.

To secure sound rails it is of supreme importance to make sound ingots. Sir Robert Hadfield, in October, 1912, presented a paper before the Iron and Steel Institute[9] which attracted wide attention and much discussion. The plan is broadly designated as using a "hot-top ingot," and many plans of obtaining such castings, other than directly following Sir Robert's, have been and are being successfully employed, notably, on ingots to be used in ordnance work. As yet hot-top ingots are not used in rail making; both encouraging and disappointing experiments have been made, but I am satisfied that it is practical and will be so proved. Based on the results obtained with other than rail ingots, its success with them is bound to come. Therefore, while I am not at this time prepared to incorporate hot-top ingots in rail specifications, the time is near when it will be entirely practical to do so. It may involve a somewhat greater first cost, to be at least partly offset by saving in scrap, but if the outlay is justified when making steel to be used in destroying life, should it not be even more so when producing metal on the soundness of which human safety will so largely depend?

The threatened shortage of high-grade manganiferous ores during the late war years led to economy in the use of metallic manganese, particularly through the more general adoption of the practice of adding it to the charges in a melted condition. The obtained results from a metallurgical point of view have been very satisfactory and specifications should insist on the practice.

The importance of the chemical composition of rail steel cannot be gainsaid. In former times, when acid Bessemer steel was used for rails, its composition was restricted to a large extent by the character of the original ores, it was controlled with considerable ease by the condition of the process itself. Heat after heat of fairly uniform composition was produced and, in short, the desired aim for the various ingredients

[9] *Jnl.* Iron and Steel Inst. (1912) **86**, 11.

of the steel was readily obtained. But basic open-hearth steel manufacture is fraught with far more difficulty, for, while the initial composition of the charge does not exert so much influence, the attainment of the desired composition for the heat is dependent on so many constantly varying conditions that a wider latitude in the range for some of the different elements must be provided. The time will come when this will not be necessary and the practice of using molten additions for recarburizing enables operations on a much more consistent basis with respect to furnishing steel of the analysis desired.

The most harmful physical characteristic of rail steel is, of course, brittleness. The low amount of phosphorus present in open-hearth steel renders protection against the dangers from that element and permits freer use of carbon as a hardening element. Experience has pretty well demonstrated that the safe upper limit for carbon in rail steel is approximately 0.75 per cent.; and accepted practice has allowed a working range of thirteen points, with the result that the lower limit of carbon has been most frequently specified as 0.59 per cent., but in some cases 0.62 or 0.63 per cent.; speaking, of course, of the almost universally accepted practice of having the steel contain not over 0.04 per cent. phosphorus. The problem of always making basic open-hearth steel within this range is difficult and frequently heats are cast slightly outside the limits specified. There is a much greater commercial opportunity for manufacturers to find other than rail uses for steel containing the lower amounts of carbon than when containing the higher amounts; therefore, they naturally aim to have the carbon content nearer the low side of the permitted range. Hence the larger percentage of heats is on the low side of the allowed carbon. In repeated cases, "off heats" are produced which fail to be within the prescribed limits of the particular specification to which the rails are being rolled, but are entirely applicable to another rail specification for which the mill has orders. This leads to the ingots from such heats being temporarily discarded, later to be reheated and rolled on the other specification. This means reheating and rolling of cold ingots, which is invariably accompanied by an increased production of second-quality rails, largely augmented by the lack of proper control and treatment in the soaking pits.

The various railway systems of the country are now using as their standard practice about the same weight of rails. There are some variations of sections, but I do not see any serious objection to the adoption of a general or common chemical specification. We know that hundredths of percentages are very delicate variations and, especially in relation to carbon, variations may occur in the results obtained by two different chemists, particularly if working under modern manufacturing pressure. If a common carbon, silicon, manganese, and phosphorus specification prevailed, the work of the open-hearth steel melter would be

simplified, the cold-ingot trouble eliminated, and better general results obtained.

As a means of lessening the tendency on the part of the manufacturers to protect themselves by keeping the carbon content of the metal toward the lower range, it would be well to accept not over an agreed upon small per cent. of heats slightly higher in carbon than the specifications. These rails could be specially marked and shipped, so that the roads would have them by themselves and use them in places for which they would be well adapted. I think that there would be but a very small percentage of such rails produced, but the very fact that the danger of rejection from over high carbon would be eliminated would cause the steel makers to feel less restricted in their procedure. In other words, it would allow a certain elasticity which would, undoubtedly, work to the advantage of both producers and consumers; of course, for acceptance, the high-carbon steel must have successfully passed the prescribed physical tests.

Another development of great interest is the triplex processes, both on the Bessemer open-hearth electric and the Bessemer and double open-hearth plans. These are being worked out in practice and, I believe, will soon be adding much toward the production of more regular and, hence, better steel, and, of course, also better rails.

I present the above points as my views, believing that investigation and discussion will prove their importance and practicability.

DISCUSSION

HENRY G. MARTIN,* Chicago Heights, Ill. (written discussion†).—The object in taking three or more ladle tests is to secure a sample that will show the average composition of the heat, not primarily to investigate the lack of uniformity existing; however, the assumption that the samples are not of uniform composition remains. In other words, we have three samples of steel of different composition, the particles of which are not subject to extreme comminution so as to mix intimately, and yield an average homogeneous sample, if placed in one container. Therefore if so mixed and the chemist weighs a portion for analysis, he does not get equal parts of each of the original samples, but, if the original samples be kept separate until they reach the balance, he can easily weigh an equal portion from each, combining the three weighings for his analysis, if it is inexpedient to run separate determinations. I have followed this practice for probably 20 yr. and find it the best means to avoid discrepancies in laboratory results. As it is, there is always difficulty in securing a homogeneous sample from one drill hole only, if proper consideration has not been given to shape and cooling of the test piece.

* Railway Steel-Spring Co. † Received Oct. 6, 1919.

G. B. WATERHOUSE,* Buffalo, N. Y. (written discussion†).—One of the most essential features of rail manufacture is the production of rails that will give good service and be free from failures. To this end rails are made to careful specifications and their manufacture is controlled by rigid inspection on the part of both the producing mill and the railroad or its representative, such as the Hunt special inspection referred to in the paper.

Captain Hunt refers to the crescent-shape base failures and describes them as ruptures in the rail flanges extending from the outside inward toward the web. This is hardly correct and may be misleading. It has been shown, over and over again, that the base breaks originating in seams invariably start at the center of the rail base, and run longitudinally in varying lengths from 3 to 12 in. and sometimes longer. Most of them are from 5 to 8 in. long and then curve outward to the edge of the flange. The split at the center of the base may exist for an indefinite time and not be discovered until the break shows on top of the flange or the piece of flange breaks off. The stresses that cause the break are explained in two ways: (1) by the eccentric bearing of the rail base, that is taking contact at one side or edge and bringing excessive strain at the center of the base which starts fracture in any seam that may be present at that point; (2) the split may be due to cross strain in the base where this member is thrown into compression, as happens where a pair of trucks are on opposite sides of the cross tie or rail-base bearing, and the head is put in tension.

A careful study of "split base breaks" shows that fully 95 per cent. occur on the tie bearing, which is further evidence that the above described causes for the initial break are probably correct. These split base breaks are almost invariably progressive, but always start at the center of the rail-base running longitudinally.

A decided advance in rail production was the invention of the deseaming machine of the Lackawanna Steel Co., referred to by Captain Hunt. It was developed in 1913 and put into use Jan. 14, 1914, deseaming the bottom of the base only. On July 13, the company began to deseam both head and base and the machine has been in constant operation ever since. The amount of rails deseamed is approximately 1,000,000 tons. The machine was thoroughly described by Captain Hunt in a paper read before the American Society of Mechanical Engineers, December, 1914, but a brief description here may be of interest.

The process consists essentially of milling off sufficient of the surface of the head and base of the rail bar, while hot and during the rolling, to secure freedom from surface defects. Not only are the seams eliminated, but the softer partly decarburized surface also, so that the rolls in the remaining passes do their work on the true higher carbon steel of the rail

* Metallurgical Engineer, Lackawanna Steel Co. † Received Sept. 23, 1919.

bar. The rail bar, with head down, is forced between two saw disks by a pair of driven pinch rolls, adjustable to bars of various sizes and having guides for the top, bottom, and sides of the bar. Adjustment at the saws is made for a cut $\frac{1}{8}$ in. or, at the most $\frac{3}{16}$ in., deep. A second set of pinch-driven rolls, on the delivery side, helps to force the bar against the cut of the saw teeth. Here, a second set of guides helps to hold the bar rigidly and firmly during passage through the saw. From the second set of pinch rolls, the bar goes along the table to the finishing rolls and receives five finishing passes. The hot rail bars enter the machine at a speed of about 350 ft. per min., are slowed down by the cutting operation to about 80 ft. per min. and on leaving the saw rapidly pick up speed until they enter the finishing stand at about 500 ft. per min. Both saw disks are approximately 8 in. wide and are belt driven from motors. They are 5 ft. diameter, with V-shaped teeth of $\frac{3}{4}$ in. pitch. The peripheral speed is 25,000 ft. per min. The teeth stand up well, and the saws mill at least 30,000 tons of rails without dressing.

The main object of this process is the production of rails with the top of the head and the bottom of the base free from seams and surface defects, thus exposing to the wheel pressures solid clean metal free from decarburized weak surface material. The improvement in head surface is of special importance in regard to wear and flow of metal. The deseaming gives the user the benefit of hard homogeneous metal at the top surface of the rail head in the initial use, which will not break down or crush over the corner of the head, as often takes place with rails that have not been deseamed. This is most desirable as it pertains to the running and service; but the improvement in the base, as a safeguard against the danger of breakage, is far more important.

It is probable that the split base failure has existed to some extent ever since T rails were put into use, but it first received special attention after A. S. C. E. Standard Sections were adopted and tie plates became a part of standard track constructions. The A. S. C. E. sections were revised into the A. R. A. sections having a thicker base. This, no doubt, reduced base failures; the use of open-hearth steel also helped a great deal. None of these revisions, however, have completely overcome base failures. The deseaming process has been very successful in this respect.

The statistics for rail failures, mentioned by Captain Hunt, give information on this point. This is not all the tonnage made, but is all the tonnage on which there is information. There were only two total base failures in the 112,936 tons of Lackawanna open-hearth rails rolled in 1914; only four in the 112,397 tons rolled in 1915; four in the 141,966 tons rolled in 1916; and none so far reported from the 97,558 tons rolled in 1917. The statistics available to date are to Oct. 31, 1917. The deseaming process was started January, 1914. Out of the 219,376 tons rolled in 1913, which were not deseamed, sixty-two base failures have been

reported to date. In other words, out of approximately 500,000 tons of rails reported on only ten base failures have developed, which is a very striking result. The results with five other mills over the same space of time is 249 base failures out of approximately 3,000,000 tons of rails.

H. T. DOUGLAS, JR.,* Chicago, Ill.—In the fall of 1915, at the suggestion of Captain Hunt, we accepted a carload of rails, for purely experimental purposes, that had not been subjected to the cold straightening process.

He selected a carload of this rail which was placed in our southbound main line track, about 5 mi. south of Bloomington, Ill., where the trains attain high velocities. The trackmen were not advised that this rail was different from the others they were receiving at the time, and they did not discover any difference in it. This rail was applied in the winter of 1915. If any of you would like to inspect these rails, we should be pleased to show them to you. The track was put up on our standard section, 12 in. of crushed rock. I saw this track and critically examined the rail two days ago: first, on the ground, and, afterward, on a high-speed train. If you were to see the high merits of this rail, it would appeal to you; we are very much pleased with the results and hope that after a while we may use no other rail. In the last shipment of rails this year, we requested Captain Hunt to send a carload of this rail, which is now on our main line tracks. The track is in beautiful alignment and surface, and I don't think it is possible for a train to ride more smoothly than the one I observed moving at 60 mi. an hour over this test rail.

ROBERT W. HUNT (author's reply to discussion†).—I am glad that Mr. Martin has given the Institute the benefit of his experience in dealing with ladle-test analyses. As I have stated, the whole matter of ladle-test ingots is now receiving the attention and study of a committee of the American Society for Testing Materials, in conjunction with the U. S. Bureau of Standards.

Mr. Waterhouse probably is right in correcting my statement about the crescent-shaped fractures. While my description might lead to the impression that I held that the rupture began from the outside and extended inward toward the web, I also stated that, as an almost invariable rule, there was a longitudinal seam in the bottom of the rail flange at the point of fracture, which was directly under the web of the rail. I certainly did not mean that the rupture began at the outside and extended inward, or I would not have emphasized the important part played by the longitudinal seam. I thank Mr. Waterhouse for making clear my somewhat imperfect description of the breaks, and am also glad that he has given us the detailed statistics of results obtained from the deseaming operation.

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*Chief Engineer, Chicago & Alton R. R. †Received Jan. 22, 1920.

Determining Gases in Steel and the Deoxidation of Steel*

BY J. R. CAIN,† B. S., WASHINGTON, D. C.

(Chicago Meeting, September, 1919)

IN every process for making steel there are one or more stages where the metal is exposed to gas of one kind or another. Thus, in the open-hearth furnace, the carbon dioxide and water vapor in the products of combustion are oxidizing toward the ferrous and manganous oxides in the slag, tending to convert these to higher oxides; the higher oxides then oxidize the steel, so that indirectly the products of combustion are oxidizing toward the steel. In the Bessemer process, there is direct contact of air and steel; also, the water vapor in the air blown into the converter is continually decomposed by the molten steel, yielding hydrogen and ferrous oxide, both of which are in part taken up by the metal. In the electric-furnace processes, there is always more or less air in contact with the slag covering the metal, and this air and the gaseous reaction products from it are dissolved in the slag to some extent and thus transmitted in part to the steel. In the crucible processes, there is always entrapped air in the charges and the crucibles are more or less permeable to the gaseous products of combustion from the furnace.

In all the processes of manufacture, the molten metal is exposed to air while tapping into ladles and ingot molds and has an opportunity to saturate itself with air or with gaseous or solid products generated by chemical reactions between air and steel. Sieverts[1] has shown that molten metals require but a few seconds to saturate themselves with a surrounding gas.

No one familiar with the operations at a steel plant needs to be reminded of the important role that gases play at all stages. The familiar "boil" of open-hearth furnaces, due to evolution of carbon monoxide when iron ore is added to the molten steel; the spectacular operation of the Bessemer converter brought about entirely by a gaseous reagent; the almost magic effect of small quantities of silicon, manganese, or aluminum added to the molten metal in ladle or furnace in quickly preventing bubbling and evolution of gases; the unwelcome phenomena of blowholes in ingots that have not been properly manufactured—these

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* Report of work at the U. S. Bureau of Standards.

† Chief, Section of Chemical Metallurgy, U. S. Bureau of Standards.

[1] *Zeit. Elektrochem*, (1910) **16**, 707.

are everyday occurrences and constant reminders of the great importance of gases in steel. The classical work of Bessemer is familiar to all metallurgists, who know that only after he found means for removing the spongy structure due to gases did Bessemer steel become of commercial importance.

Nor is the effect of gases in the metallurgy of steel limited to the melting and casting operations. Nearly all the further stages of manufacture are affected in one way or another by gases. Indeed, the metallurgy of iron and steel occupies almost a special position in this regard, due to the considerable solubility of gases in the metal and to the opportunities for the generation of gases by reaction of the carbon in the metal with oxygen or water vapor from the always present air. In rolling and forging ingots, blowholes caused by gases may cause much trouble if they do not weld, and such trouble is magnified when the billets are rolled into sheets, for the area of the blowholes is correspondingly enlarged and a few unwelded blowholes may cause the loss of many sheets. In the sheet mills, the occlusion or adsorption of hydrogen when the metal is pickled or cleaned in acid can cause much trouble if the gases are not removed by annealing. The role of gases in case-hardening operations has only come to be recognized in recent years, so that now one of the most generally applicable methods of case-hardening uses a gaseous substance to carry the carbon into the metal. The effects of nitrogen and oxides in electric and oxyacetylene welding operations are also recognized. Practical steel manufacturers seem to have always felt that the gas content of various kinds of steel was of great importance and many have shown a tendency to roughly grade steels according to their probable freedom from gases. Thus, they would rank crucible and electric-furnace steels first, because these ought to be most free from gases, next are the open-hearth steels, and finally Bessemer steel. There has nearly always been more or less assumption that gases could only be harmful. Very little work has been done to prove whether or not such an assumption is always correct, at least in so far as concerns the ordinary physical properties of steel.

Effect of Gases on Physical Properties of Iron and Steel

From a less technical but more scientific standpoint, there is abundant evidence of the remarkable effect exerted on the properties of iron by the small quantities of gases that it contains. Thus, hydrogen in electrolytic iron has been shown by Roberts-Austen,[2] Burgess, Yensen,[3] Cain, and Schramm and Cleaves,[4] to be the cause of extreme hardness

[2] *Proc.* Inst. Mech. Engrs. (1899) 42.

[3] Univ. of Ill. Engng. Ex. Sta. *Bulls.* 72 and 83.

[4] U. S. Bureau of Standards *Sci. Paper* 266 (1916).

and brittleness. Braune[5] has shown that nitrogen causes brittleness, and affects the thermal transformation points and the magnetic properties. Yensen has shown, in a series of remarkable tests, that the magnetic properties of iron melted in a vacuum are much better than of iron melted in ordinary ways.

Previous Researches on Methods for Determining Gases in Steel

In view of these important effects of gases on manufacturing processes and on the physical properties of steel, it is not surprising that much work has been done on analytical methods for determining the gas content of steel. But, although many such researches have been carried on since the pioneer work of Fremy, Deville and Troost, Cailletet,[6] and of Graham[7] in this field (1861–1866), it is remarkable that today there are no analytical methods in general use in steel works for determining the gaseous constituents of steel. The exact chemical control of the phosphorus, silicon, sulfur, carbon, and manganese in steel manufacturing processes is now almost universal, but constituents such as nitrogen and oxygen, which may be of equal importance in many cases, are seldom or never determined. In fact, the almost vital process of deoxidation, which really ought to be called degasification, has hardly advanced at all since the first days it was practised, largely, probably, because there have been no adequate chemical methods for ascertaining the mechanism or the efficiency of the process. The explanation of this is comparatively simple and is to be found in the great experimental difficulties in obtaining reliable results for the gas content of steels. Such experimental difficulties have been emphasized in one way or another by nearly every one who has worked on the subject. Thus, Parry in describing his work on gases in steel says, "It was, however, found that the experiments entailed great difficulties and even now after nearly six years work the author is of the opinion that this important subject is yet far from being exhausted." The fact that the work of these earlier investigators has not led to the devising of analytical methods that have attained general metallurgical importance does not indicate any lack of skill or ability. Indeed, considering their want of electrically heated furnaces, reliable mechanical vacuum pumps, vacuum-tight combustion tubes, and suitable gas analysis apparatus—all of which are at the command of present day investigators—it is surprising that so much has been done.

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[5] *Rev. de Met., Extraits* (1907) 834.

[6] *Comptes Rendus* (1861) **52**, 323; (1863) **57**, 965; (1865) **61**, 850.

[7] *Proc.* Royal Soc. London (1866) **15**, 502; (1867) **16**, 422; (1869) **17**, 219; *Jnl.* Chem. Soc. (1867) **20**, 257.

In view of the very evident importance of the general subject of gases in steel and with particular reference to its bearing on deoxidation processes, the Bureau of Standards has started some systematic investigations in this field and it is the purpose of this paper to state the scope of these, what is sought to be accomplished, and the general results thus far attained.

Before dealing with these Bureau investigations, however, it seems best first to refer, very briefly, to the general classes of methods that have thus far been developed, giving for each method its limitations, as they appear to the writer; second, to give a summary of the most important facts established by past investigators as to the composition and quantity of the gases in steel and physical laws relating to absorption of such gases by the metal; third, to state some of the important questions not yet answered by the researches of other workers. In this way it is possible to consider the Bureau investigations in proper perspective.

A quite satisfactory review of certain phases of the subject has been given recently by Alleman and Darlington[8] and by Sir Robert Hadfield.[9] For the present paper, the methods used by others for examining the gas content of steel have been arranged arbitrarily in the order of experimental difficulty.

The simplest method is to drill the metal under water, oil, or mercury and collect the gases liberated during this operation, as has been done by Stead and Muller. There is some disagreement in the findings of those who have used this method and it is obvious that only a part of the gas can be thus liberated. This work may be said to have established qualitatively the presence mainly of hydrogen, carbon monoxide, and nitrogen, with a little carbon dioxide and oxygen. The actual or relative amounts of gases indicated by this method can hardly be said to have any significance. It is of some interest, however, that any part of the gas can be liberated by mechanical work in the cold. Later investigators have found that rolling also decreases the gas content of steel. These facts seem to indicate that at least a part of the total gas content is held imprisoned between crystals or aggregate of crystals.

Probably the next method in apparent simplicity of execution is the Ledebur method for determining oxygen or oxides, which consists in reducing these by hydrogen and collecting and weighing the water thus formed. This method can determine only certain of the combined forms of oxygen likely to be present in the metal and hence is of quite limited use in studying the total gas content. The details of these limitations are given in one of the Bureau publications referred to later in this paper.

Methods have been used that liberate the gases by dissolving the

[8] *Jnl. Frank. Inst.* (Jan.–Apr., 1918) **185**, 161.
[9] *Trans.* Faraday Soc. (1918).

metal, or by using solvents that react with the gases to form products that can be accurately determined. Goutal attempted to determine carbon monoxide and carbon dioxide by dissolving the steel in copper-potassium-chloride solution and collecting the gases liberated but, as will be shown later, Bureau investigations have revealed fundamental defects in this procedure. Nitrogen has been determined in steel by Allen, and others, by dissolving the metal in dilute hydrochloric or sulfuric acid. Under such conditions, ammonia is formed from the nitrogen and this is determined by standard methods. It seems quite probable that this class of methods yields only the nitrogen combined in certain ways, as detailed later. It is not likely that any imprisoned or dissolved uncombined nitrogen can be so determined. Nevertheless the results by such methods can be obtained by simple manipulation and, as will be shown later, they will probably yield information of value when properly interpreted.

Many workers have heated steel, in vacuo, to various temperatures below its fusion point, collecting and analyzing the gases evolved. This has been a favorite method and much more generally used than the fusion methods next to be described, inasmuch as the technique is simpler. Graham, Parry, Troost and Hautefeuille, Belloc, Boudouard, Baker, Sieverts, Charpy and Bonnerot have used this procedure. The results, in general, show that very long periods of heating are required for the complete evolution of gas at any given temperature. Apparently, too, when the gas evolution has ceased at a given temperature, it starts again as soon as the temperature is raised. All this indicates that the gas recovered by heating solid steel in vacuo can be only a part of the total. In other words, these methods are somewhat in advance of the methods whereby the gas is liberated by drilling under water or oil, but still they are not perfect. Unfortunately, in some of this work there were probably large errors due to reactions between the steel and the materials on which it was supported, which reactions liberated gas. In spite of these objections the combined work along these lines has contributed valuable information in confirming the qualitative analyses of the gases obtained by the drilling methods.

Comparatively few investigators have extracted the gases from fused steel. Here the experimental difficulties are really very great, since it is necessary to work at temperatures not below 1500° C., and some means must be devised to maintain a good vacuum at this or higher temperatures. It is only recently that this has become possible. No porcelain or other refractory tubes made will meet this test, consequently it is not possible to heat externally; instead, the heat must be developed internally by such devices as the Arsem furnace, the induction electric furnace, or some form of cathode-ray heating; or current may be passed directly through the piece to be fused. This last method, employed by Austin, has the

disadvantage that the metal cannot be kept fused long enough for complete liberation of all its gases. Besides Austin, Bessemer, Cailletet, Troost and Hautefeuille, Goerens and Paquet, Sieverts, Walker and Patrick, and Baraduc-Müller also extracted gases from fused metal.

The work of Goerens and Paquet[10] is of special interest in that they used a device for obviating the difficulties of holding vacua at high temperatures. By mixing the steel with antimony and tin, an alloy melting below 1200° was obtained and they could thus carry on their work at temperatures where quartz or porcelain tubes are satisfactory. Their investigation seems to be the first systematic attempt to establish a relationship between gas content of the metal before and after addition of deoxidizer. The results in this regard are rather limited in number, however, and there are objections because of the employment of very small samples (only 2 or 3 gm.) and of relatively very large weights of alloying materials. Nevertheless, they have made a substantial contribution to our knowledge on these points and in addition have shown that rolling the metal decreases the gas content and that there is often considerable segregation of gas in steel. The general results of these methods confirm in a qualitative way those obtained by the methods of drilling and of heating the solid metal in vacuo.

Summary of Results Obtained by Others

Without attempting to go into details, the net result of the previous work on gases in steel may be considered to have established these points:

The gases concerned are mainly hydrogen and carbon monoxide with considerable nitrogen and a little carbon dioxide.

Certain or all of these gases may be present in three well recognized forms: (1) Imprisoned in pores or blowholes or between crystal boundaries; (2) in solid solution; (3) as compounds in solution or suspension (oxides, nitrides).

The gases may be given up in varying degrees when the metal is heated in vacuo to various temperatures or when it is drilled or when it is otherwise worked hot or cold.

The amounts of gas per unit weight or volume of metal reported by previous investigations are in general too discordant to permit safe conclusions to be drawn.

It will be noted that the facts thus far enumerated have qualitative value only. However, the extensive work of Sieverts, which was quite thorough and related to other metals than iron, as well as to alloys, and which was carried out apparently with all the precautions necessary, adds some results of a more quantitative nature:

The solubility of gases in metals increases with the temperature and

[10] *Ferrum* (1915) **12**, 57, 73.

generally increases much more rapidly after the melting point is passed.

The solubility of gases in alloys is sometimes the combined solubility of the constituents of the alloy and sometimes follows other laws according to the constitution of the alloy.

The solubility of gases in metals is proportional to the square root of the pressure.

Another investigation that gave quantitative results that could not be affected by any of the errors usually characteristic of work on fused steel was that of Baraduc-Müller,[11] which is especially mentioned at this point for this reason and also because of the vivid impression it gives as to the very large quantities of gas involved and the indication of their probable effect on physical properties. Baraduc-Müller made ingots weighing over 1000 lb. (453 kg.) by casting in special molds placed under evacuated steel tanks and pumped off the gases evolved as the steel solidified; these gases were then analyzed. The method he used could not give more gas than was present, for there was no chance for the molten metal to come into contact with anything with which it could react and generate gas. On the contrary, all the gas could not have been evolved, because the ingot solidified too soon. Nevertheless, 1895.3 l. of gas were evolved in the time required for an ingot weighing 1232 lb. to cool to ordinary temperature. The average analyses of the gases from a certain ingot showed 52.2 per cent. hydrogen, 30.5 per cent. carbon monoxide, 12.7 per cent. nitrogen, the rest being oxygen and carbon dioxide. Physical tests of metal not freed from gas compared with similar tests of the same metal freed from gas showed, for the latter, about 10 per cent. superiority in tensile strength, 30 per cent. increase in elastic limit, but about 20 per cent. decrease in elongation. Unfortunately, however, the data on the physical tests are not as conclusive as might be desired, since the two sets of specimens were not in entirely comparable condition as to mechanical work done upon them.

Investigations at the Bureau of Standards

After having attempted to present the work of others in respect to the more notable results, it is worth while to consider what has not been accomplished or what yet remains to be done. An adequate investigation ought to give the following results:

1. Eliminate faults of existing analytical methods or establish new ones that are simple, reliable, and quick enough to be generally applied in steel works.

2. Establish the solubility of the individual gases in molten and solid steels of industrial compositions and the effect of these gases on physical properties.

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[11] Iron and Steel Inst., *Carnegie Schol. Mem.* (1914) **6**, 216.

3. Show the specific effect of the commonly used deoxidizers on gas content of steels and demonstrate the theory controlling their use; that is, whether they increase the solvent power of the metal for gases, or expel or combine with the gases present, and how much deoxidizer is necessary for a given gas content in undeoxidized metal.

4. Indicate desirable new deoxidizers or deoxidizing procedures for securing the minimum gas content in industrial steels.

5. Show relations between the gas content and the industrial methods of pouring or treating ingots; that is, the comparative gas contents of the same steel cast in various shaped molds with and without sink-head, cast in vacuo, cast by compression processes, etc.

6. Give full information as to the effect of mechanical work on gas content.

By contrasting the requirements of this program with the statements just preceding as to what has been accomplished, it will be noted that although the net result of preceding work, with few exceptions, has been of qualitative value, the program calls for quantitative methods and results. Here the preceding work either fails on all the terms enumerated in the program or else too little work has been done or it has been done by questionable methods. Specifically, the work thus far done fails because:

1. It is well known that no methods for determining total gas content of steels are now in use at works.

2. Even the extended investigation of Sieverts on the solubility of gases in metals covered only a few steels and did not pretend to be complete in this regard; no attention at all was paid by him to the effect of gases on the ordinary physical properties.

3. The principal studies on steel deoxidizers have been conducted by Pourcel, Brinell, Boylston, and by Goerens and Pacquet. In only the latter two investigations were gas analyses made and not enough were made nor were the experiments, generally speaking, extensive enough to justify the formulation of any theory of deoxidation.

4. Nothing at all has been done to discover whether new deoxidizing combinations give lower gas content than those commonly used.

5. No analyses, reliable or otherwise, are in existence giving a comparison of gas contents of the same steel poured by the different methods of casting ingots. Such a comparison between vacuum cast ingots, as made by Baraduc-Müller, and ingots made by the Harmet, Whitworth, or similar compression processes would be of very great interest, for the vacuum-cast ingots would give the minimum possible content of gas and the compression processes the maximum possible; in neither case could there be blowholes to complicate matters.

6. The work of Goerens and Paquet, which laid no claims to completeness in this regard, is all we have as to effect of mechanical work on gas content.

The work being done at the Bureau of Standards is part of the program of the Chemical Metallurgical Section, whose more general functions relate to the preparation of metals and alloys of extreme purity or sometimes of commercially pure grade. This work on gases has been in progress in a limited way for the past few years and has gradually oriented itself according to the program just outlined. Doubtless the first impression of this program is that it involves so much research that there is no prospect of completing it in any reasonable time. This is true in a certain degree, but on the other hand, the completion of each phase of such a program will probably bring results of considerable scientific and technical interest, thus stimulating research on the part of other investigators.

It was partly with this idea in mind and partly because of the comparative simplicity of this part of the problem, that the first Bureau investigations have dealt with analytical methods for determining gases in steel. The aim has been to examine critically some of the simpler methods in order: to remedy any defects; to make them more rapid, if possible; and to form some idea of the metallurgical significance of the results obtained by them.

The final object of this phase of the work, which is not at all complete as yet, is to devise a few simple methods which it is hoped will really be useful to the steel manufacturer in controlling his deoxidation processes. To do this, it is not sufficient merely to test these methods on a few samples of unknown metallurgical history. Probably the surest way to attain success is to test the method on heats of steel deoxidized under conditions chosen with a view to securing definite variations in gas content. For this purpose a small experimental furnace of 500 lb. (226 kg.) capacity is being built at the Bureau for melting steel under conditions similar to those prevailing in open-hearth furnaces, or the conditions can be varied as desired. An electric-arc melting furnace of 500 lb. capacity is also available for this work.

Some of the Bureau investigations now to be presented in abstract have already appeared as Bureau publications and will be printed in full elsewhere; the other Bureau investigations herein mentioned will also soon be available as Bureau publications or in the technical press.

Three researches on the Ledebur method have been completed.[12] "A Critical Study of the Ledebur Method for Determining Oxygen in Steel" shows that very special precautions are necessary in obtaining hydrogen of sufficient purity and combustion tubes sufficiently non-

[12] J. R. Cain and Earl Pettijohn: A Critical Study of the Ledebur Method for Determining Oxygen in Steel. U. S. Bureau of Standards *Tech. Paper* 118 (1919).

J. R. Cain and Leon Adler: Equilibria Between Carbon, Ferrous Oxide and Hydrogen in Relation to the Ledebur Method for Oxygen in Steel. U. S. Bureau of Standards *Sci. Paper* 350.

J. R. Cain and Earl Pettijohn: Determinations of Oxygen by the Ledebur Method on Steels Deoxidized in Various Ways. U. S. Bureau of Standards *Sci. Paper* 546.

porous for this method, and working directions are given on these points. It is shown that chips for these analyses have to be taken with very great care in order to prevent surface oxidation during cutting, and a practical method is described for cutting under oil, together with the precautions necessary to remove the oil. Unless chips are thus prepared, errors as great as 150 per cent. may result. On the other hand, if chips are thicker than 0.1 mm. not all the oxygen determinable by the Ledebur method is obtained. A special electrically heated furnace was devised for the work and is illustrated in the original paper. The oxides that may be present in steel and determinable by the Ledebur method are shown to be FeO (or lower oxides of iron), CO_2, and part of the CO; NiO, Cu_2O, and WO_3 if present would also be determined, but these seem unlikely constituents of steel. On the other hand, MnO, SiO_2, Al_2O_3, TiO_2, and Cr_2O_3 are not determined. Even FeO is not determined if it is combined as silicate, as seems likely to be the case. Thus, the real application of the Ledebur method seems to be extremely limited, notwithstanding the fact that it has been used probably more than any of the other methods proposed for determining gases in steel.

In order to throw light on an overlooked source of error in this method, a research was conducted to ascertain how much of the reduction of iron oxide is effected by hydrogen. Ledebur assumed that he could determine any oxygen in steel present as ferrous or lower oxides of iron by causing the oxygen to combine with the hydrogen that was passed over the chips. He seems to have overlooked the fact that in steel there is also another reducing agent present, namely, carbon, and that there might be a partition of reducing action between carbon and hydrogen.

The paper on "Equilibria Between Carbon, Ferrous Oxide and Hydrogen in Relation to the Ledebur Method for Oxygen in Steel" shows that, at best, probably only 75 per cent. of the reduction is effected by hydrogen and that consequently the Ledebur method under most favorable conditions fails to determine the other 25 per cent., thus demonstrating still further the lack of value of this procedure.

Finally, the Ledebur method was applied to the determination of oxygen in several heats of acid Bessemer and Tropenas converter steel made during deoxidizer investigations at the Watertown Arsenal and the Bethlehem Steel Co.'s plant at Sparrows Pt., Md. Ferromanganese, spiegel, ferrosilicon, aluminum, ferrotitanium, and carbon-free ferrotitanium were used to deoxidize parts of the same heat, or similar heats of steel, under quite a variety of conditions.

The paper on "Determinations of Oxygen by the Ledebur Method on Steels Deoxidized in Various Ways" gives the details of these deoxidation processes and the analyses for oxygen made at the Bureau by the revised Ledebur method. The oxygen contents of the various steels were all practically identical, although the steels themselves showed

considerable differences as to certain physical properties, and particularly as to freedom from blowholes, so that here again the Ledebur method proved of little value.

Goutal's method for determining carbon monoxide and dioxide in steel consists in dissolving the metal in copper-potassium-chloride solution and collecting and determining the liberated gases. This seemed a very simple procedure and one that would be useful at works; consequently it was investigated at the Bureau. An account of this work is given in "A Critical Study of the Goutal Method for Determining Carbon Monoxide and Carbon Dioxide in Steel."[18] This paper shows that the carbon dioxide and carbon monoxide yielded by the Goutal method do not come from gases present in the steel, but are generated by oxidizing action of the cupric chloride on the easily oxidizible carbides present. Hence, this very attractive method is worse than useless.

The Allen method for determining nitrogen in steel has been thoroughly tested at the Bureau. This method depends on solution of the metal in hydrochloric or sulfuric acid (which converts at least part of the nitrogen into ammonia), making the solution strongly alkaline with sodium hydroxide and distilling off the ammonia, which is then determined by titration against standard acid. There are a number of variations of the procedure, and many refinements are necessary. Most of these have been tried out in the Bureau work.

As already stated, it is quite probable that the Allen method determines certain nitrides only (probably nitrides of manganese); some other nitrides likely to be present in steels, free nitrogen, and conceivably nitrogen in some forms of combination with carbon are not converted to ammonia by reaction of the acid on the steel and would not be shown by the Allen method or its modifications. It is believed, however, that if the steel were heated to high enough temperature, practically all these compounds would give up their nitrogen as such, so that if a method were available for accurately determining this, it would be possible to get reliable figures not only for the total nitrogen content of steels, but also, by measuring the nitrogen given off at different temperatures, it would be known where each compound dissociates and thus the nitrides in steel could be identified with more or less certainty. Such a direct method for determining nitrogen has been devised in our laboratories. Although tests of it are not considered entirely complete, enough has been done to show that the principles are correct and that a direct determination of nitrogen in a mixture, regardless of other gases present, can be made with comparative ease. This is far more accurate and reliable than where this gas is determined by difference, as has been done heretofore. The principle of this method is to absorb the nitrogen at reduced pressure by

[18] J. R. Cain and Earl Pettijohn: U. S. Bureau of Standards *Tech. Paper* 126 (1919).

calcium vapor, which combines with it to form calcium nitride. The calcium nitride is then decomposed by acid, forming ammonia, and the rest of the determination is carried out by the Allen method. This work on nitrogen will be published from time to time and it is believed that simple methods for this constituent will soon be available. The evidence showing important effects of this gas on steel is fairly concordant, and the use of reliable methods for determining it may reveal some interesting additional data.

A limited amount of work has been done by the Bureau on the proper conditions for determining carbon monoxide in gases obtained from steel. The methods employed depend on the use of copper oxide, or nickel or cobalt oxides, as catalyzers for differential or selective oxidation of this gas to carbon dioxide. A selective oxidation method is desirable since the gas mixture exhausted from steel may under certain conditions contain hydrocarbons, and the catalyzer for the carbon-monoxide oxidation should not cause oxidation of these hydrocarbons since high results would then be obtained. The hydrocarbons in the gas mixture are due to secondary reactions and are not believed to be present in the steel itself. This work will be published at some future time.

Preliminary work on methods for determining the total gas content of steels is well advanced. The method chosen involves fusion of the steel in vacuo, since it is very evident that only in this way can accurate results be expected. Two methods are being worked out. One involves heating the steel by an induction electric furnace where the temperature can be carried as high as desired, since the heat is generated in the metal and does not have to be conducted through the walls of the tube that has to hold the vacuum. The other method permits determinations to be carried out at temperatures where ordinary refractory tubes are gas-tight. The first method requires expensive appliances that might not be generally used, but the second will probably not be open to this objection. Both methods will be compared on a larger number of steels before any results are published. Special materials for crucibles, means for obtaining these free from gases, and special mechanical features are embodied in this work. It is believed that these methods now being developed for total gas content are the ones that will be of special value in the study of deoxidation processes, and since such studies give promise of being of value in the industry, special attention will be given to trying to perfect a method or methods that will be used at the works.

As has been already stated, an electric-arc melting furnace and a small open-hearth furnace, each of about 500 lb. capacity, will be available at the Bureau for the experiments on deoxidation of steel which are to be controlled by the gas analyses. While such method of control is one of the ultimate ends of this research, it is evident that much more is involved, for it is also desired if possible to make better steel than has been made by

deoxidation processes hitherto in use. This means that temperatures, slags, and ingot practice shall be taken into consideration and, finally, that the ingots shall not only be satisfactory specimens as to blowholes, piping, and segregation, but that their behavior on rolling and forging shall be satisfactory and that they give superior physical tests. Fortunately the Bureau now possesses a satisfactory rolling, forging, and heat-treating equipment all on a scale proportional to the melting furnace, so that the necessary equipment for such a research is at hand.

Two other researches now in progress in the Chemical Metallurgical Section of the Bureau have to do with the deoxidation of steel. One is being carried out in coöperation with the National Research Council and a group of eleven laboratories representing industrial, university, and testing laboratories. This research has for its ultimate object the selection of deoxidizing alloys not hitherto used. The criterion used for the suitability of the alloy is the fusibility and viscosity of the slag formed by its use as a deoxidizer. To test this, various binary and ternary mixtures of oxides of aluminum, silicon, titanium, and manganese were made up and their fusibility determined, eliminating from consideration any slag melting above 1500° C., since such could not separate readily from steel. In this way it is possible to limit in a simple manner the number of alloys worth investigating and a still further limitation can be made by viscosity measurements. When a group of deoxidizing alloys has been selected in this way they will be made and tried in the furnaces at the Bureau. This investigation was started during the war, when there was still some apprehension regarding our domestic supply of manganese and was intended to afford some relief by cutting down the amount of manganese required for deoxidation and for encouraging the use of substitutes, but regardless of this question the work seems well worth completing.

The same incentive initiated another investigation now well along toward completion in one phase. The purpose of this was to define more exactly the functions of manganese in steel, so as to give necessary information for revision of specifications, to diminish in some cases the per cent. of this constituent. As is well known, manganese is added to steel as a deoxidizer, as desulfurizer, and for conferring physical properties. The investigation is intended to define the amounts necessary for each purpose. The effect on physical properties has been reinvestigated by the preparation of a series of iron-carbon and iron-carbon-manganese alloys of graded percentages made from electrolytic iron and pure carbon and manganese. Most of these alloys have been made and rolled and are now to be heat-treated and physically tested. The other two phases of the investigation have not been started.

It is believed that this paper should make clear the far-reaching nature of any investigation that will attempt to give the information needed on

the effect of gases on steel and deoxidation problems. All operations common to the metallurgy of steel are involved: melting, deoxidation, ingot practice, casting, forging, rolling, heat-treating, galvanizing, and many others. The centralized metallurgical equipment of the Bureau and its well-equipped chemical and physical laboratories should produce valuable researches, provided necessary assistance is secured from the industries. Such contributions would doubtless stimulate research elsewhere along such lines and substantial progress could be made. A few companies have already volunteered for limited assistance and it is believed others will follow as the work goes on.

DISCUSSION

ARTHUR F. BRAID,* New York, N. Y. (written discussion†).—I have read Mr. Cain's paper with a very great deal of interest. The subject is one that is worth a great deal of thought and research at this time.

It is a well-known fact that the strong affinity of steel or iron for oxygen is perhaps the main detriment to sound production, and, as the author states, it is indeed remarkable that today there are no analytical methods in general use in steel works for determining the gaseous constituents of steel.

The research work now being carried out by the chemical metallurgical section of the Bureau of Standards on the deoxidation of steel is very interesting to me, and I had the pleasure of discussing this matter with Mr. Cain in Washington some time ago. I agree with the author, that the suitability of an alloy for deoxidizing purposes is based on the fusibility and viscosity or specific gravity of the oxides formed. The composition of the well-known 25 per cent. carbon-free ferrotitanium was no haphazard guess; the composition of this alloy is based on the fusibility and low specific gravity of the double oxide of titanium and aluminum that is formed when carbon-free ferrotitanium is added to steel.

ALLERTON S. CUSHMAN, Middletown, Ohio (written discussion‡).—This subject is one to which I have given much thought and study and which I have frequently discussed informally with Mr. Cain and others. I have long been convinced that the gaseous elements, either combined or entrained (occluded), in steel play a much more important role than is ordinarily admitted. This is probably due to the difficulties encountered in the accurate quantitative estimation of gases, some of which Mr. Cain has referred to, as well as to the fact that, considered from the weight-

* Metallurgical Engineer, Metal and Thermit Corpn. † Received Sept. 3, 1919.
‡ Received Sept. 19, 1919.

percentage basis, gas content is measurable only in hundredths or even thousandths of 1 per cent. Volumetrically, however, the results obtained by modern research are startling, to say the least. Baraduc-Müller,[14] as cited by Mr. Cain, pumped out of 1000-lb. ingots of basic Bessemer steel 2100 l. of gas. or about 75 cu. ft., at normal temperature and pressure, per ton of steel. In other words, we are to understand that Bessemer steel normally carries about fifteen times its own volume of gas. According to Baraduc-Müller's table of results, which is given herewith, carbon monoxide, nitrogen, and hydrogen are important constituents of Bessemer steel, the latter gas being undoubtedly formed by the decomposition of the moisture water in the atmosphere during the process of blowing.

	Per Gross Volume, Liters	Per Cent.	Per Ton of Steel, Liters
Carbon dioxide	42.2	3.6	76.7
Oxygen	10.6	0.9	19.3
Carbon monoxide	352.2	30.5	640.3
Hydrogen	604.3	52.2	1098.7
Methane	2.4	0.2	4.3
Nitrogen	147.7	12.7	268.5

Baraduc-Müller also says:[15]

In the case of hydrogen, it is seen that the 5150 cu. m. of air injected into the converter contained, owing to the 5.672 gm. of water per cu. m., 29,210.8 gm. of water capable of yielding, by complete dissociation, 3245.6 gm. of hydrogen, corresponding, at about 15° and under a pressure of 760 mm., with a volume of 36,058.6 liters.

Now with 1098.7 l. of hydrogen given per ton of steel, measured at the ordinary temperature, the entire cast must have contained, at a given moment, 13,860.3 l. of hydrogen. It results, therefore, that the maximum amount of hydrogen fixed, at any rate momentarily, by the steel in some form or other, dissolved or in combination, must have been

$$\frac{13{,}860.5 \times 100}{36{,}058} = 38.5 \text{ per cent.}$$

of the total volume of the available hydrogen.

This throws an interesting insight into the extreme solubility of the gases, and in particular of the hydrogen, in liquid steel at a high temperature. It remains to ascertain if these gases are actually in solution or in combination, and also what is left of these gases in the steels at the moment of solidification in the ordinary condition of manufacture.

Austin, whose work has also been referred to by Mr. Cain, investigated the gas content of samples of mild open-hearth steel and obtained about 1 c.c. of gas per gm. of steel, equivalent to about 35 cu. ft. per ton.

[14] Iron and Steel Inst., *Carnegie Schol. Mem.* (1914) 6, 216.
[15] *Op. cit.*, 227.

The analysis of Austin's gas is as follows: Carbon dioxide, 7.7 per cent.; carbon monoxide, 18.4 per cent.; hydrogen, 59.1 per cent.; nitrogen, 14.8 per cent.

It is interesting to compare these results with those obtained by Baraduc-Müller who cooled basic Bessemer steel in a vacuum, and obtained 75 cu. ft. of gas per ton of metal, while Austin obtained in the case of the open-hearth steel 35 cu. ft. per ton; or approximately 7.5 cu. ft. of gas to each cubic foot of metal. In the first case, however, it is presumable that some part of the gas collected would have escaped into the air, had it been allowed to cool in a normal manner. Austin, however, started with a finished bar of cold steel, and it is difficult to escape the conclusion that all the gas evolved was actually held in some form or another in the body of the metal. That the gas was present in blowholes is impossible, when we consider the volume relations that follow from his experiments. Austin collected gas which, calculated to a tonnage basis, amounted, as has been shown, to about 35 cu. ft. per ton of metal. This is a most extraordinary conclusion and one that will perhaps be unwelcome information to many steel makers. Fortunately or unfortunately, however, depending on the point of view, there is quite a mass of evidence to show that commercial steels as made and sold contain up to ten times their volume of gas.

In 1871–74, Parry examined a large number of commercial steels by heating them in evacuated porcelain tubes, but he obtained such high yields of gas that no one would believe the accuracy of his results. Later, Belloc, Boudouard, Baker, Charpy, and Goerens examined a number of steels and irons by heating in vacuum and obtained results which show that a wide variety of steels yield anywhere from six to ten volumes of gas per unit volume of metal. While many of these researches may be so open to criticism that general conclusions based on them should be conservative, they point in one direction, viz., that the gas content of steels is much higher than has been generally supposed, and introduces an important factor when the soundness and physical character of any given type of metal has to be studied.

The question as to whether the gaseous impurities that are combined with iron or solid impurities in the form of hydrides, nitrides, and carbonyl compounds are also removed by heating or melting steel and iron in vacuo, is still unanswered. Baker suggests that the most probable, and at the same time the simplest, explanation is that the gases are simply imprisoned in the pores of the steel, and when the steel is reheated in vacuo they diffuse out.

If the physical properties of steels of various types are affected by the nature and quantity of the occluded gases they contained, it seems fair to the writer to inquire whether it may not be true that these factors exert an important influence on the much discussed question of corrosion.

resistance in relation to the chemical constitution of iron and steel. The alleged good or bad effect of minute differences in the percentage composition in relation to carbon, manganese, sulfur, silicon, and especially copper, has been so much discussed by a great number of investigators that there is little left to be said or claimed in regard to the influence of these solid constituents. The effect of gas content has, however, been curiously overlooked in the discussion of corrosion problems heretofore, and yet it is probable that this one factor is the most important of all with relation to all the commercial metals, no matter whether we are considering a steel sheet or a brass or bronze condenser tube.

Apparently steel partakes of the nature of a microscopic sponge capable of holding enormous quantities by volume of gas. We need more accurate information in regard to the effect of this gas content on the physical characteristics of steels, including comparative resistance to corrosion but one common-sense conclusion appears to stand out prominently, and that is that the highest purity as represented by the most efficient degasification is a consideration of the highest importance in the manufacture of all high-quality metals.

Mr. Cain has pointed out that the total gas content of steels subdivides into three separate classes. I think that this point should be thoroughly understood by all students of iron and steel, and in order to simplify our nomenclature I would suggest that we refer to alpha, beta, and gamma gas in all discussion of this subject. Alpha gas is simply the atmosphere contained in pockets, blowholes, pipes, and seams contained within the body of the metal. Beta gas is molecularly entrained or occluded in a state that is defined by physical chemistry as solid solution, or, in other words, it is held submicroscopically. Gamma gas is combined or fixed gas, as in oxides, nitrides, hydrides, and carbonyl compounds, which may be evenly dissolved in the metal or segregated in its structure.

We can now discuss the limitations of analytical and gas extraction methods as applied to metals. The Ledebur oxygen determination, as pointed out by Mr. Cain, determines all of the α and β oxygen plus a part of the γ. Of course, any oxygen that may be present combined in non-reducible oxides, such as manganese oxide, is not determined by the method. Nevertheless, the Ledebur method yields very useful results in the control of deoxidation. The Allen method for nitrogen determines the γ nitrogen only and Mr. Cain suspects only a part of this; a new method for estimating total nitrogen is being worked out at the Bureau of Standards. It is to be hoped that if this method is too difficult and costly to be used in commercial control of degasification, it will at least yield data that will enable us to apply approximate corrections to the easily carried out Allen method. Practical everyday methods for the determination of carbon monoxide and total gas content would be of the greatest val[redacted] steel-works research laboratory.

There are numerous problems connected with the relations of gases to steel that are not yet solved, and in regard to which there are few data available. If, for instance, steel is annealed in hydrogen at certain temperatures, it becomes soft and loses elasticity, while some authorities have claimed that hydrogen is absorbed or combined. If steel is pickled in acid, it absorbs hydrogen and becomes more brittle. No explanation of these divergent effects of hydrogen on steel has come to my attention as yet. Again, if steel is annealed in an atmosphere of ammonia gas, it absorbs large quantities of nitrogen, and perhaps some hydrogen, while the material is rendered so brittle that it will break like glass. It is probable that in these experiments the gases associate themselves with the iron in either the β or γ conditions, or in a combination of these conditions.

In a long series of analyses for nitrogen carried out by the research department of The American Rolling Mill Co. by a method similar to that of Allen (which, as we have seen, determines only γ nitrogen), it has been observed that the nitrogen content of sheet steel that has been exposed to atmospheric corrosion for extended periods is invariably higher than in samples cut from the same sheets before exposure. This observation needs further confirmation and study, but if it is found to be a fact, it is certainly an extraordinary phenomenon, for it would then appear that iron is able to fix the nitrogen of the air, although in minute quantities, at atmospheric temperatures and pressures. We have also collected data to show that Bessemer steels unusually high in γ nitrogen corrode with extraordinary rapidity. This may be the explanation of the well-known fact that, other things being equal, Bessemer steel suffers from corrosion to a much greater extent than open-hearth metal. In general, purity and efficient degasification are marked characteristics of rust-resisting metals. The collecting of the data on which the above general observations are based has not yet been concluded, so the data are reserved for future publication and discussion.

J. S. UNGER,* Pittsburgh, Pa. (written discussion†).—Those investigators who have tried to determine oxygen in steel by reduction in hydrogen or oxides in slags by solution methods have been surprised to find that at times their results showed great variations on the same samples. When one considers that oxygen and nitrogen are blown into the blast furnace and cupola and the reaction is intensely oxidizing, then powerfully reducing, it is difficult to understand how oxygen can be found in the pig iron, especially, in the presence of high percentages of such reducing agents as carbon, manganese, and silicon.

Carbon-monoxide gas is used as a cementing agent in certain processes. Steam passed over red-hot iron is decomposed leaving an atmosphere of

* Central Research Bureau. † Received Oct. 6, [redacted].

hydrogen, while part of the oxygen unites with the iron to form a durable protective coating used in the arts. A mixture of natural gas and ammonia gas passed over red-hot iron is an extremely rapid cementing agent. It is probable that cyanides and nitrides are formed as intermediate products, but the end product is carburized steel. From these data, it would appear that gases are not necessarily injurious.

Results of tests showing pig iron containing more oxygen than Bessemer steel, which is made by a rapid oxidation process, have been published. Other tests show that electric steels made under reducing slags have a higher oxygen content than basic open-hearth steels made by oxidizing methods. Some published results show steels of superior quality but of higher nitrogen content than other steels containing less nitrogen. Such evidence is conflicting. Shimer and Kichline[16] showed how difficult it was to prepare over-oxidized steel. Iron and many other metals absorb gases at ordinary temperatures. Much has been written of the beneficial or injurious effects of the elements purposely added or found in steel. Much is based on opinion, not on facts.

It would appear to me that accurate data on the effects of various gases on the physical and mechanical properties of steel are of equal, if not greater, importance than the development of the best methods for the determination of gases. These data can only be obtained by making steels containing large amounts of oxygen, nitrogen, hydrogen, or their compounds, and making a careful study of their effect on the properties of the steel. If it is shown that a certain gas, or gases, has particularly beneficial or deleterious effects, then the best method of determining this gas should be studied.

SAMUEL L. HOYT,* Nela Park, Cleveland, Ohio (written discussion†). The work that Mr. Cain is doing on gases in steel should have a highly important bearing on investigational work in connection with steel-making practice. We found, for example, in our work in connection with the use of manganese in open-hearth practice that we needed expressions for the manganese efficiency and the "condition" of the steel. To get these figures we assumed that we would need to know the amount of FeO at different stages of the heat and the amounts of the gases present. This required simultaneous analyses of our samples for FeO and the gases and Mr. Cain seems to be the only investigator who has considered the analytical work from this point of view. That he has done so is very encouraging and it is to be hoped that his work will solve the mysteries now surrounding this neglected phase of steel metallurgy.

Those interested in the Ledebur method for use in connection with the efficiency of manganese as a "deoxidizing" agent will be disappointed

[16] *Trans.* (1913) **47**, 436.

* Metallurgical Engineer, Experimental Engineering Laboratory.

† Received Oct. 30, 1919.

to learn that more recent work at the Bureau indicates that Ledebur determinations are not reliable. The principal source of error seems to be due to the partial reduction of the ferrous oxide by the carbon present in the steel. However, it seems like an unnecessary criticism of the Ledebur method to state (p. 195) that it proved of little value because it showed little differences in oxygen for a series of steels treated with various reagents and showing variations in physical properties and freedom from blowholes. One would hardly expect to find appreciable differences in Ledebur oxygen between "deoxidized" steels; and, even so, the amount of Ledebur oxygen in treated steel should bear no particular relationship to the blowholes present. It is well known, for example, that casting steel or forging steel must be treated with a "degasifier," such as silicon or aluminum, in addition to the "deoxidizer," manganese. The manganese is certainly capable of reducing ferrous oxide but it does not eliminate blowholes. On the other hand, both Oberhoffer and Schmitz[17] in their recent work with this method show a drop in the "Ledebur oxygen" with deoxidation by ferromanganese. If this is substantiated by work at the Bureau, it may prove to be possible to check analytically the efficiency of the manganese addition and the degree of oxidation subsequent to this addition.

On p. 197, it is proposed to use, as a criterion of the suitability of an alloy for deoxidation purposes, the fusibility and viscosity of the slag formed during such use, and to test this by experiments on various oxides. There are several important objections to the use of this test. The alloys used in the oxidation process, and for which the alloys in question are actually used, are generally assumed: (1) to react with oxides in the steel and form, as reaction products, oxides of the constituents of the alloy added, and (2) to act in some way, as yet not understood, to inhibit the evolution of hydrogen and nitrogen. Evidently only the first reaction is considered, and the efficiency of the addition in the second reaction would be missed by the test. Certain alloys are used principally on account of their efficiency in the second reaction and would be entirely too costly to use for the first, for which manganese seems to be eminently suited.

The assumption is made, tacitly, that the constituents of the addition oxidize in the same ratio as they occur in the alloy, and that the oxidation products unite. This assumption seems quite logical but its validity will not be assured until the reaction products are identified. In our work on manganese we felt that a big advantage in using manganese-silicon alloys would be obtained because a silicate slag would form, but we could not advance that as a positive advantage because identification of the reaction product was lacking. So important did it seem to identify reaction products that the writer has gone into the matter

[17] *Stahl u. Eisen* (1918).

through a grant from the Carnegie Scholarship Fund (Iron and Steel Institute).

There is the possibility that a certain amount of iron oxide is simply dissolved in the reaction product. This action would depend considerably on the concentration of the reducing agent. This may or may not be a minor point but it should be considered when determining the criterion to use for deoxidizers. It must also be borne in mind that the reaction products form during deoxidation in the presence of a large excess of iron, while synthetic slags form under different conditions. Again the necessity of identifying reaction products is evident.

The gases present in steel are classified into three groups (p. 190). The classification as advanced involves considerations of the homogeneous equilibrium of the steel because it would be difficult to distinguish between gases that are in solution as uncombined gases and as compounds. It might be advisable, for the present, to place gases in solution and gaseous compounds in solution in the second group.

T. D. YENSEN, Pittsburgh, Pa.—Our knowledge of gases in metals, their causes, their effect on the physical properties, and methods of controlling them is very fragmentary. It is sufficient to enable the steel makers to produce high-grade iron and steel according to present-day standards, but great improvements can undoubtedly be made. Furthermore, there is too much variation in the quality of the steel, even if it is made by one mill and according to one method. This is shown by the large factors of safety considered necessary for structural steel and the allowable variation in magnetic properties for electrical steel. The subject is an extremely complicated one.

Mr. Cain was kind enough to refer to some of my work on the magnetic properties of iron and iron alloys and the remarkable improvement obtained by melting the iron in vacuo. I think that the idea prevails that by such treatment the iron can be completely degasified. While this may be true in certain cases, I could exhibit small ingots of iron and iron alloys that have been kept molten for several hours under an absolute pressure of less than 2 mm. mercury and allowed to cool under this same low pressure down to room temperature, that are full of blowholes. Certain combinations are more susceptible to blowholes than others, for instance certain iron-nickel alloys, but in all these cases even a prolonged vacuum treatment is not sufficient to prevent the formation—or liberation—of gases upon cooling. Far more effective than vacuum treatment in preventing these blowholes is the use of some agent. For some materials, silicon can be used; for others, aluminium, manganese, titanium, etc., all of them strong deoxidizers. It is, therefore, probable that the blowholes are caused by the reaction of two or more elements that are held in solution, to a small extent, by the molten metal, but are caused to react after the metal has solidified, causing the blowholes.

This explanation is old, but needs verification. It is one of the numerous questions that await Mr. Cain and his associates, and to answer them they will have to use refinements in methods of analysis hitherto unheard of.

With Mr. Cain's method of analysis, an accuracy of ±0.01 per cent. may be realized. In the investigations of the physical properties of iron and iron-alloys, it has been found that carbon, even in minute quantities, influences these properties to a very large extent. Quantities that have hitherto been regarded as traces have been found to be sufficient to change certain properties, 100 or 200 per cent. An accuracy of ±0.01 per cent. is therefore not regarded as sufficiently accurate in such an important investigation as that undertaken by Mr. Cain. At the Westinghouse Research Laboratory, we have been working on this problem of carbon determination for a long time, and we believe that we have developed a method whereby an accuracy of better than ±0.001 per cent. can be attained. It is a modified combustion method, in which the CO_2 from the sample is frozen out by liquid air and then allowed to evaporate into a known evacuated volume; the resulting pressure is then noted. It is believed to be a simpler method than Mr. Cain's, but requires the use of liquid air and, preferably, vacuum. However, both of these agencies are now considered essentials in modern research laboratories and should cause no objection. The determination of CO_2 can readily be made with an accuracy of ±0.0001 per cent., using a 2 gm. sample. The chief errors will be caused by the blank of the combustion tube. I feel certain that this method will meet with favor wherever it is tried.

J. A. Upperle, Middletown, O.—We have found that the gas content of sheet metal plays an important part in producing a good enameled product. The vitreous enameling industry demands a sheet that will enamel perfectly, one that is free from occluded gas and will not blister. Pure iron has been found to be ideal for enameling on account of its uniformity in chemical analysis and physical characteristics, the latter quality being conducive to the production of a flat product free from warping.

We have also found that the gas contained in sheet metal affects it from a corrosion standpoint. We have made a study of iron that has had a long life, and without exception we have always found the nitrogen content very low. On the other hand, we have analyzed steel which has had a short life, under normal atmospheric conditions, and we have invariably found such steel high in nitrogen. We have studied Bessemer steel sheets in which the elements except nitrogen were about the same; the nitrogen varied from 0.007 to 0.021 per cent., and when 26-gage uncoated sheets were exposed to the atmosphere at Middletown, Ohio, the high-nitrogen sheets failed in 13 mo., while those low in nitrogen lasted several times as long.

The Ledebur method for determining oxygen is recognized as having its limitations, but where manganese and silicon are low, such as in pure iron, we have found the method of great help in maintaining a uniform product. We have modified the Ledebur method so that we determine the oxygen, carbon monoxide, and carbon dioxide in one operation. The oxygen is absorbed as water in phosphoric anhydride, the carbon dioxide in barium hydroxide, and the carbon monoxide is oxidized to carbon dioxide with the use of iodine pentoxide and then absorbed in barium hydroxide.

While the modified Ledebur method may not give exact results, the information which the method yields is of great value in the production of sheet metal having a low gas content.

The hydrogen content of metal has been investigated, most hydrogen being introduced into the metal by pickling or by electrolysis. We have had some electrolytic iron so high in hydrogen that the metal was as brittle as glass and contained so much hydrogen that it could be ignited with a flame. We also found that heating metal in hydrogen gas in a silica tube would make such metal extremely soft and ductile, and even though the original metal contained considerable combined hydrogen, such hydrogen was eliminated when the metal was heated in hydrogen. In other words, we found it impossible to make hydrogen and iron combine by heating iron in an hydrogen atmosphere.

THE CHAIRMAN (J. W. RICHARDS, So. Bethlehem, Pa.).—I wish to agree with Mr. Yensen's statement of the very small amounts that will cause the gas. We forget that the gases in the liquid steel are produced at a temperature of 1600° C.; therefore, if they are also under a tension of 2 mm., their volume will be 2600 times what it would be at atmospheric pressure and at 0° C. so that under such a condition as mentioned by Mr. Yensen, 0.001 per cent. of gas by weight would have a volume 135 times that of the steel, at the temperature of the steel bath. This explains why steel will boil out of a crucible if a suction pump is put on it when the steel is liquid. It also explains why quantities of gas undeterminable by analytical methods can form very unsound ingots and steel castings.

H. M. RYDER, Pittsburgh, Pa.—I have been very much interested in reading of the plans of Mr. Cain. His scheme seems to be, principally, a study of the effect of different treatments of steel on the gas content, while some work that I have been doing has been principally on the effect of the gases on the physical and chemical properties of the metals. The two methods are, of course, complimentary and overlap, the difference being largely a matter of emphasis, and together, if completely worked out, should certainly offer a satisfactory solution to a problem which has been worked on spasmodically for fifty years or more.

One phase of the problem, which is important from the standpoint of

certain manufacturers who work with high-grade steels, which Mr. Cain has not specifically mentioned, although it has been implied by him, is the determination of the nature of the retention of these gases in steel. Large quantities of carbon monoxide which, upon heating, may be removed from steel are retained as absorbed gas and also in the form of carbon and a metallic oxide. We can hardly imagine hydrogen being held in a gaseous form in iron at an equivalent pressure of 200 atmospheres. Again, it has been ascertained that a considerable quantity of carbon dioxide, removable from steel in a furnace, is due to grease or some other organic substance on the specimen, even though this latter has been carefully washed in boiling water. I believe that an understanding of this point of view regarding the gases in metals does much toward the simplifying of the problem from the point of view of the user of steel as well as from the producers of this product.

Another phase of this problem, which Mr. Cain suggests in his discussion, but does not mention specifically, is a use of the knowledge of gases in steel, and their peculiarities, to indicate chemical or physical changes which may take place, as well as the nature of these changes. I believe that, of the gases which may be removed from steel, a relatively small percentage is in the uncombined state in the solid metal and hope to be able to present some satisfactory evidence to this effect shortly. If this is the case, the evolution of a gas, which has been combined, upon elevating the temperature of the specimen, must be the evidence of some reaction, or physical change, and therefore a clue for study. A curve I presented before the American Electrochemical Society at its Southern Meeting last year,[18] which showed the sudden evolution of certain gases at the A_2 point for that steel, may serve to illustrate the type of information it is possible to obtain toward this objective.

I have, after a detailed study of this problem, great faith in its possibilities, and I was very much pleased to hear of Mr. Cain's confirmation of the intention of the Bureau of Standards to offer its facilities to the solution of a problem as broad, I believe, as has been considered in the metallurgy of steel for some years.

WILBERT J. HUFF, Pittsburg, Pa.—One of the speakers spoke of the corrosive effect of the nitrogen content of iron and steel; that point is especially interesting to the Bureau of Mines for we are carrying on some experiments on the corrosion of rifle barrels. A theory for this corrosion, enunciated some 14 or 15 years ago by a German lieutenant, attributed some of the queer effects of corrosion to the solution of acid gases in steel and the slow leaking out or diffusion, if you will, to the surface of the gun barrel. I want to disclaim any responsibility for that theory myself, because I do not believe it to be correct, but the point is an ex-

[18] *Trans.* Amer. Electrochem. Soc. (1918) **33**, 197.

tremely annoying one. Is this nitrogen, alpha gas, beta gas, or gamma gas, and does anyone have any experience that will show that gases held in solid solution will diffuse through steel at ordinary temperatures?

CHAIRMAN RICHARDS.—The presence of the gases and their deleterious effects may sometimes be counteracted by the use of proper protective agents in the manufacture of steel. It is possible that such agents as a small amount of vanadium in the ore or in the pig-iron exert a protecting influence during the manufacture, so that nitrogen, particularly, is kept out of the final product, and thus its deleterious effect is minimized. This idea of a protection from the influence of the gaseous ingredients, or the suppression of the gaseous ingredients, by proper protectives, will be an important one to the steel maker when we know more about it; it is prevention rather than cure.

J. R. CAIN.—Doctor Cushman's suggestions are very good, and some such system of nomenclature for these gases will no doubt be adopted in time. It seems to me that it would be difficult to carry out Doctor Unger's suggestion that we first saturate the metals with the gases that we are studying and see whether or not they have any effect, until there is some accurate method of determining the gas under investigation. If you are determining the effect on steel of phosphorus, sulfur, or silicon, the very first thing is to determine the amount of the element. I believe that is the only practical method for determining the effects of gases. I quite agree with him as to the discordant evidence in the literature as to the effect of gases. Probably many of the determinations are in error.

Doctor Yensen's remarks are of value, inasmuch as the Westinghouse Research Laboratories have been doing much work on this subject. As to his comment on my method for determining carbon in steel, I would say that the method is intended for works use in meeting specifications, none of which, I believe, specify the carbon content closer than 0.01 per cent. If the specifications are refined and it is found desirable to specify to 0.001 per cent., there is no question that the method could be modified to give that degree of accuracy. There is no doubt but that a few thousandths per cent. carbon when present in certain forms will have a very marked effect upon the magnetic properties of some specimens of iron.

Mr. Feild asked whether ferrous oxide is the oxide present in steel. I do not know. There is evidence that lower oxides than FeO exist, but whether these are present in iron, I cannot say.

Mr Braid's discussion brings out another criterion of a good deoxidizer: namely, an alloy that will produce a deoxidation slag of low specific gravity so that it will tend to rise to the surface of the molten metal.

Professor Hoyt is mistaken in assuming that the experimental de-

oxidation treatments used in the Bureau furnace with new deoxidizers developed by the research outlined on p. 197 will not also include a study of degasification. In fact, it is specifically stated on p. 196 that "a small open-hearth furnace . . . will be available at the Bureau for experiments on the deoxidation of steel which are to be controlled by the gas analyses." In order to simplify the choice of trial deoxidizers, certain assumptions had to be made, such as the equal rate of oxidation of both deoxidizing elements and the separation of the resultant slag as such from the metal without its dissolving iron oxide. Had all such variables been taken into consideration, it would have been impossible to get results with any reasonable amount of labor. The investigation being carried out ought to orient us somewhat concerning some of these variables and point out the direction for new researches of a more complete nature.

I feel that it was quite justifiable to expect actual differences in residual oxygen content of steels with such radically different deoxidation treatments as those described in the Bureau investigation to which Professor Hoyt refers. One would otherwise have to assume perfect deoxidation for all the steels, or that they were all incompletely deoxidized to the same degree, both of which assumptions seem very improbable, and the fact that the Ledebur method as used did not indicate any differences in these steels shows to a considerable degree its uselessness in studying deoxidation of ordinary steels. Professor Hoyt seems to make the assumption in these remarks that all the "deoxidation" in these experiments was done by the manganese (since it was first added to all the heats) and none by the added silicon, aluminum, titanium, or carbon, which is also very improbable

The comments of Professor Richards concerning the volume occupied by the gases in steel and Mr. Ryder's remarks along the same line are of great interest. It seems to me, however, very unsafe to calculate the space occupied in the steel by the gases, basing the calculation on the percentages of gases extracted by heating or melting the sample in vacuo. Doubtless much of the gas thus obtained results from the reduction or dissociation of nitrides, oxides, silicates, etc. (which occupy a relatively small volume in the steel) in contact with carbon, and probably only a small proportion of it exists in the steel uncombined.

Flaky and Woody Fractures in Nickel-steel Gun Forgings

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(New York Meeting, February, 1919)

IN connection with certain coöperative work carried on between the Ordnance Department of the U. S. Army, the U. S. Bureau of Mines, and the U. S. Geological Survey during the year 1918, it was the writers' privilege to make metallographic examinations of a great many test bars and other pieces of nickel-steel gun forgings. The material studied came from many different steel plants throughout the country and consisted both of forgings that had been accepted by the Ordnance Department and those that had failed to pass the specified physical tests. In general, these tests require a minimum tensile strength of 95,000 lb. (42,091 kg.), an elastic limit of 65,000 lb. (29,483 kg.), an 18 per cent. elongation, and a 30 per cent. reduction in area. An average chemical composition of the steel studied is 0.38 per cent. carbon and 2.9 per cent. nickel. The steel was of both basic electric and basic open-hearth manufacture. The object of the metallographic work was twofold: to assist in detecting defects in the steel and, if possible, to suggest remedies for the same; and to obtain as much information as possible in regard to the relation of the microstructure of the metal to its physical properties. The greater part of the investigation was devoted to defective metal that failed to pass the ordnance specifications. Many types and kinds of more or less serious defects thus came under observation, but most of these were due to the attempts by the manufacturer to meet the enormous demand for ordnance steel with more or less inadequate equipment, and were soon remedied.

Two defects, the so-called "flakes" or "snowflakes" and a certain "fibrous" or "woody" fracture, were encountered so frequently and were so disastrous to the steel that the greater part of the study was devoted to them. As a result much information as to their nature and their effects upon the steel and something in regard to some of the conditions that favor their development has been obtained. The purpose of the present

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paper is to present a brief description of these defects, with the thought that this information will produce a discussion that may result in the solution of a very troublesome and, in some instances, a very serious problem in the manufacture of large nickel-steel forgings. The writers request the fullest and most unreserved criticism of any part of the paper.

Location of Defects

Both woody, or fibrous, texture and flakes appear most prominently in the steel when it is broken in a direction transverse to that of elongation in forging. Either defect when present is shown to good advantage in any transverse fracture of the metal, whether pulled test bars or nicked and broken, or split, blocks.

Woody Fracture

Woody fracture is a term used by ordnance inspectors to describe the fracture of a pulled test bar that has a fibrous appearance and to a certain extent resembles the fracture of a piece of wood, such a fracture is shown in Figs. 13 and 15. In some instances, this type of fracture has a decided "platy" appearance and occasionally careful examination will reveal more or less foreign matter, largely slag, between the plates or fibers in the fracture. The foreign matter is by no means present in all, nor even in the majority of such fractures. In fact, it was only in the beginning of the greatly increased demand for ordnance steel due to the war, when the manufacturers were making such great efforts to supply ordnance material with more or less inadequate equipment, that slag and other foreign matter were present in sufficient amount to cause trouble. This was soon remedied. The woody fractures in many cases still remained even in steel that leaves little to be desired as to freedom from sulfide, slag, and similar impurities, showing clearly that such fractures are due to a condition of the steel itself and not to foreign inclusions.

Two features that appear to be characteristic of both the woody and the flaky steel are brought out when polished surfaces of the metal are etched for either macroscopic or microscopic study. Copper-bearing etch mediums, such as Rosenhain and Haughton's, Stead's, or Stead-Le Chatelier's reagents, never fail to reveal a decidedly non-homogeneous metal with large and prominent but much elongated dendrites. These features are illustrated in Figs. 27, 28, and 29. The microstructure, as brought out by any of the commonly used etch reagents, such as picric acid or amylic nitrate, closely resembles that of flaky steel and will be described in the paragraph devoted to that subject.

FIG. 1.—FULL CUP DULL GRAY FRACTURE OF BAR 16. × 3½.

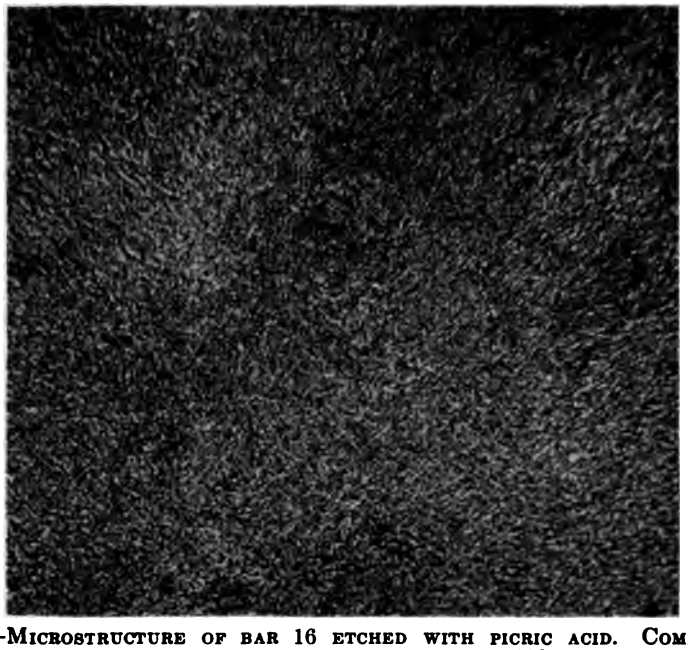


FIG. 2.—MICROSTRUCTURE OF BAR 16 ETCHED WITH PICRIC ACID. COMPLETELY GRAIN-REFINED AND SORBITIZED. × 260.

FIG. 3.—THREE-QUARTER LIPPED STAR FRACTURE, DENSE, FINE GRAY OF BAR 7. × 3½.

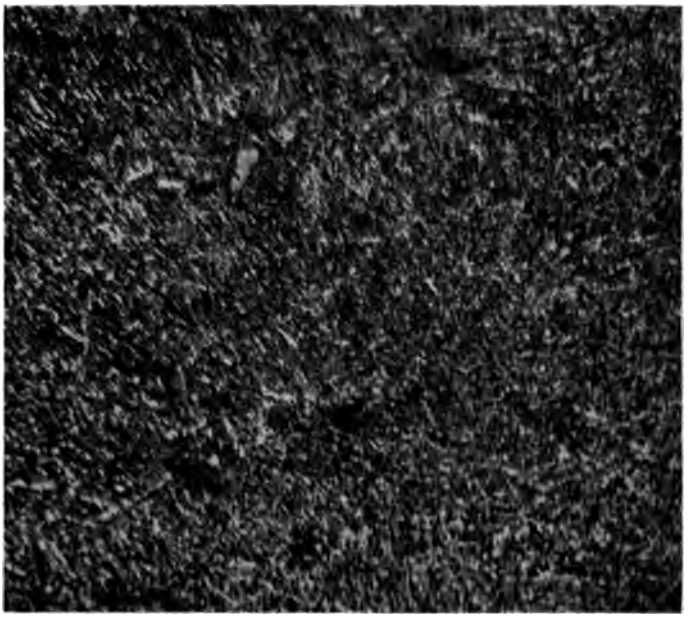


FIG. 4.—MICROSTRUCTURE OF BAR 7 ETCHED WITH PICRIC ACID. SEEMINGLY GRAIN-REFINED BUT SHOWING SOME FREE FERRITE. × 260.

FIG. 5.—FRACTURE OF BAR 20. ANGULAR. 75 PER CENT. FINE CRYSTALLINE AND 25 PER CENT. LAMINATED WITH STREAKS. × 3½.

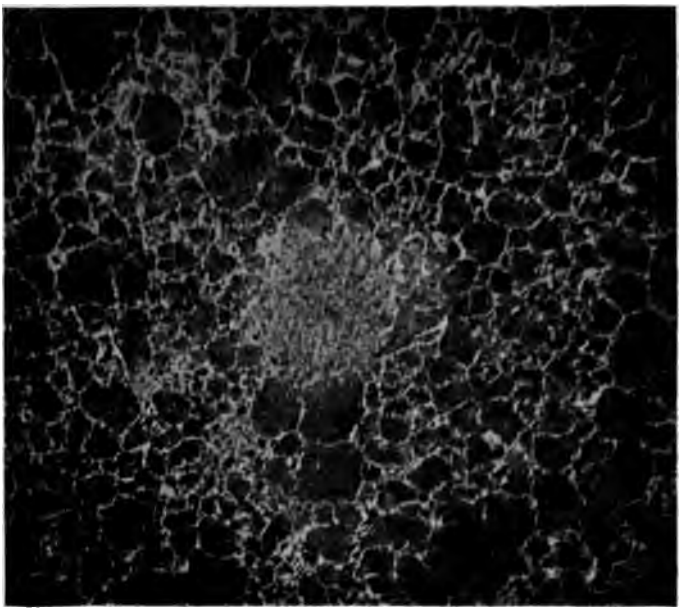


FIG. 6.—MICROSTRUCTURE OF BAR 20 ETCHED WITH PICRIC ACID. AN UNEVEN NETWORK WITH EXCESSIVE FERRITE PRECIPITATION IN SEGREGATED AREAS. HIGHER MAGNIFICATION SHOWS FERRITE AREAS TO CONTAIN NUMEROUS INCLUSIONS. × 260.

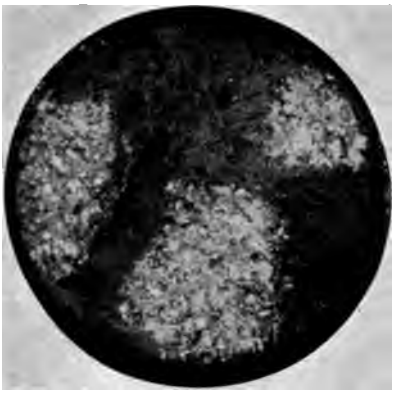


FIG. 7.—FRACTURE OF BAR 36. SILKY AND LAMINATED WITH ABOUT 50 PER CE: FLAKE. × 3½.

FIG. 8.—MICROSTRUCTURE OF BAR 36 ETCHED WITH PICRIC ACID. COARSE NI WORK OF NEEDLE-LIKE FERRITE AND SORBITE. AREAS CONTAINING CARBON SHO TO BE IN SOME INSTANCES SORBITIC AND IN OTHERS APPROACHING TRUE PEARLIT × 260.

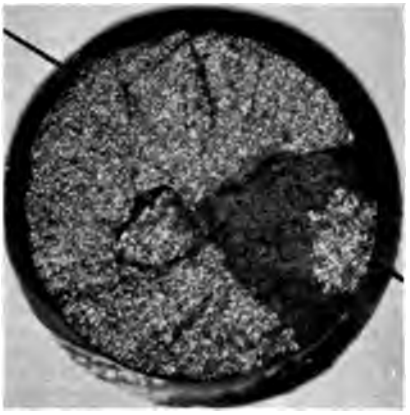


FIG. 9.—FRACTURE OF BAR 38. ABOUT 75 PER CENT. FINE GRANULAR AND 25 PER CENT. LAMINATED WITH FLAKE. × 3½.

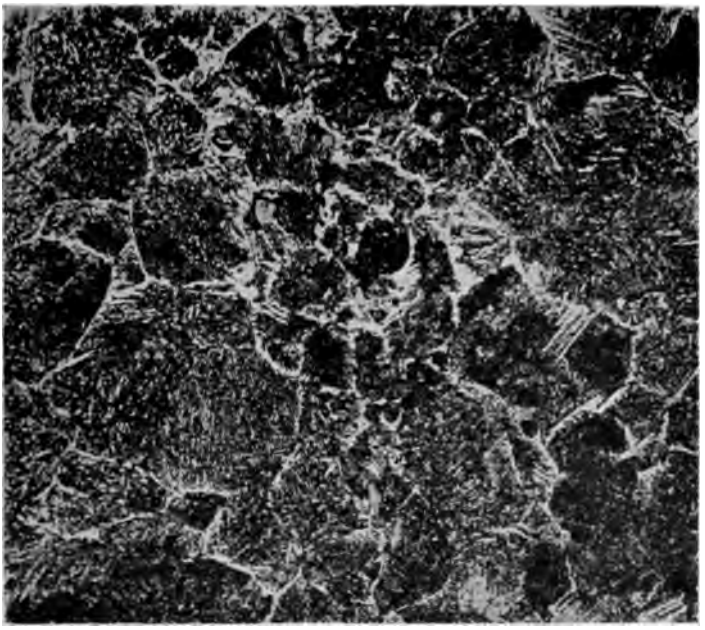


FIG. 10.—MICROSTRUCTURE OF BAR 38 ETCHED WITH PICRIC ACID. COARSE NETWORK OF NEEDLE-LIKE FERRITE AND SORBITE, LATTER IN CERTAIN AREAS CONVERTED TO LAMINATED PEARLITE. MORE FREE FERRITE SHOWN THAN IN FIG. 8. × 260.

FIG. 11.—FRACTURE OF BAR 032. ABOUT 50 PER CENT. FINE GRANULAR AND 50 PER CENT. COARSE FLAKE. × 3½.

FIG. 12.—MICROSTRUCTURE OF BAR 032 ETCHED WITH PICRIC ACID. SORBITIC AND SEEMINGLY GRAIN-REFINED WITH FERRITE ROUGHLY OUTLINING A PREVIOUS COARSE NETWORK. × 260. THIS BAR WAS HEATED 7 HR. AT 1600° AND COOLED IN AIR, HEATED 5 HR. AT 1375° AND COOLED IN FURNACE, HEATED 4 HR. AT 1475° AND QUENCHED IN OIL, DRAWN FOR 4 HR. AT 1155°, REHEATED 4 HR. AT 1475° AND QUENCHED IN WATER, DRAWN FOR 4 HR. AT 1165° AND COOLED IN FURNACE.

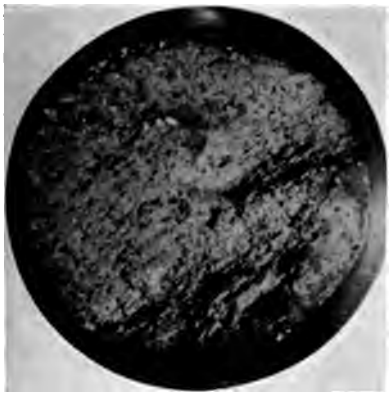


FIG. 13.—FRACTURE OF BAR 39. LAMINATED OR WOODY WITH SLAG. × 3½.

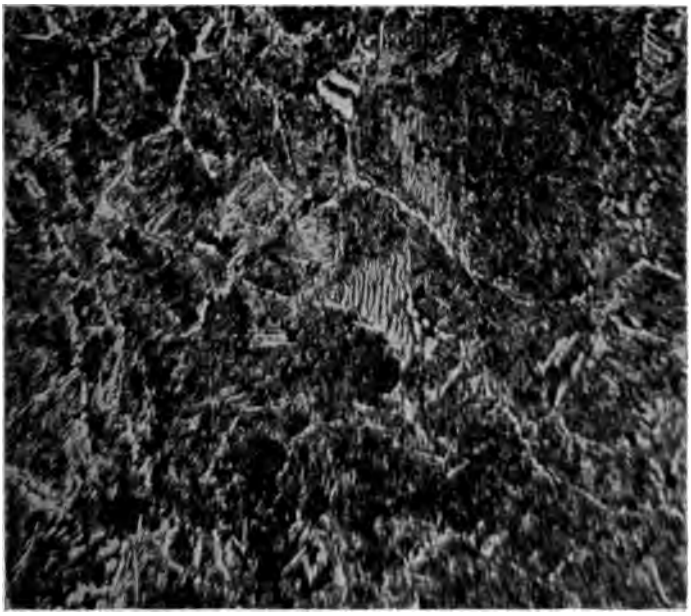


FIG. 14.—MICROSTRUCTURE OF BAR 39 ETCHED WITH PICRIC ACID. COARSE NETWORK OF NEEDLE-LIKE FERRITE AND SORBITE WITH WIDER AREAS OF FERRITE. × 260.

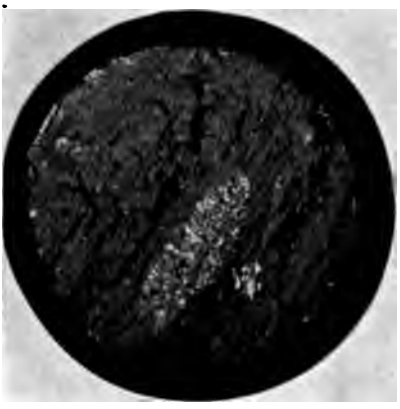


FIG. 15.—FRACTURE OF BAR 8 LAMINATED OR WOODY WITH STREAKS AND SLAG. × 3½

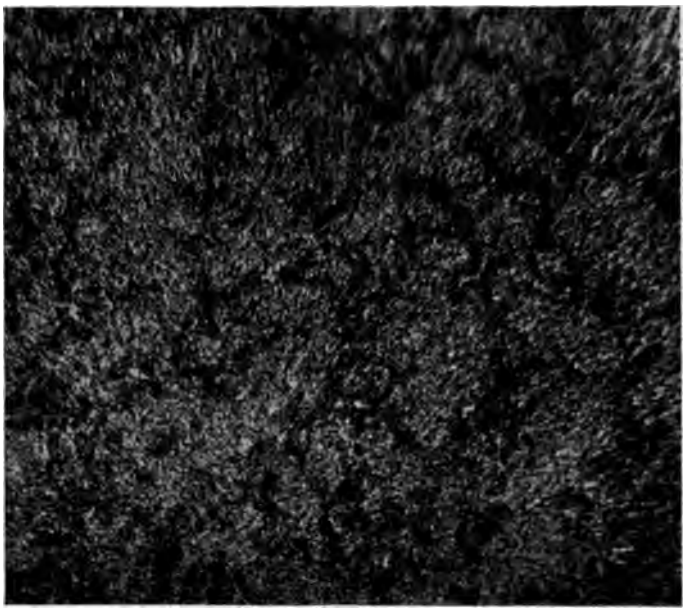


FIG. 16.—MICROSTRUCTURE OF BAR 8 ETCHED WITH PICRIC ACID. SEEMINGL GRAIN-REFINED SORBITE, BUT CLOSE EXAMINATION SHOWS DARK FUZZY LINES INCLO ING THIN FERRITE STREAKS AND ROUGHLY OUTLINING A PREVIOUS NETWORK. × 20

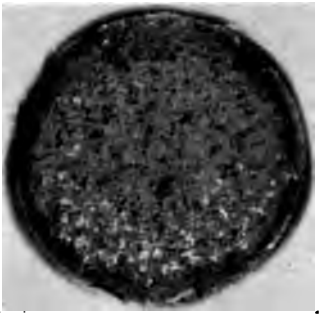


FIG. 17.—FRACTURE OF BAR 18 ONE-HALF LIP. 40 PER CENT. FINE CRYSTALLINE FINE GRANULAR AND 60 PER CENT. GRAY LAMINATED.

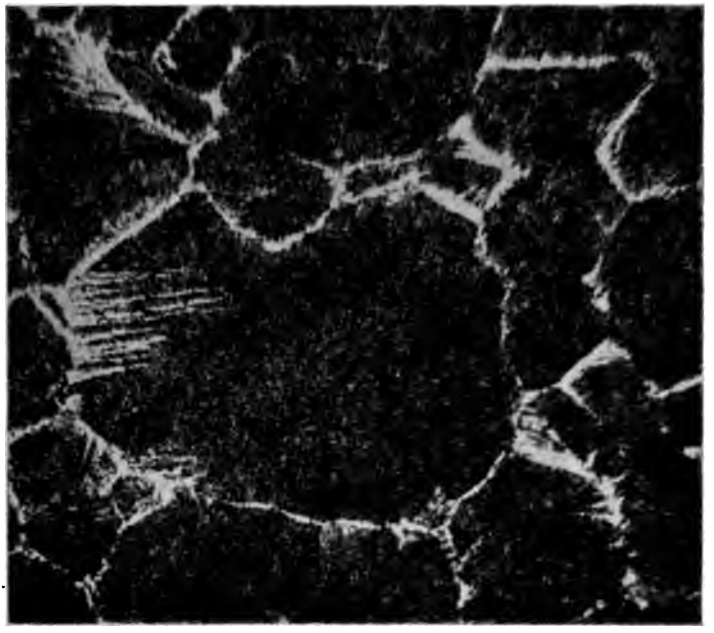


FIG. 18.—MICROSTRUCTURE OF BAR 18 ETCHED WITH PICRIC ACID. EXCEEDINGLY COARSE NETWORK OF SORBITE AND NEEDLE-LIKE FERRITE. COARSEST NETWORK EXHIBITED BY ANY SPECIMEN OF HEAT-TREATED NICKEL-STEEL FORGINGS EXAMINED IN THIS LABORATORY. × 260.

FIG. 19.—PHOTOGRAPH OF TEST PIECE SHOWING CRACKS THAT DEVELOP IN FLAK BARS AT RIGHT ANGLES TO THE DIRECTION OF PULL. × 3½.

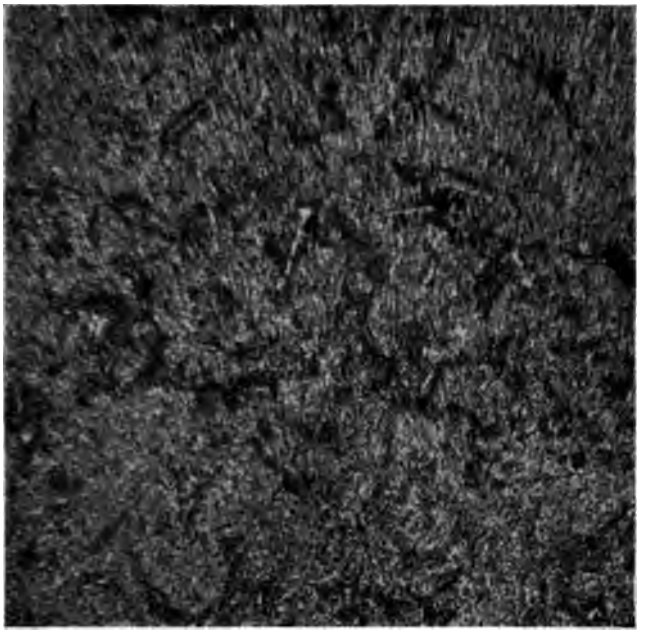


FIG. 20.—MICROSTRUCTURE OF BAR. SEEMINGLY GRAIN-REFINED, BUT PREVIOU COARSE NETWORK IS ROUGHLY OUTLINED BY FERRITE AND A CONSTITUENT WHIC ETCHES DARKER THAN SORBITE. COMPARE WITH FIGS. 12 AND 16.

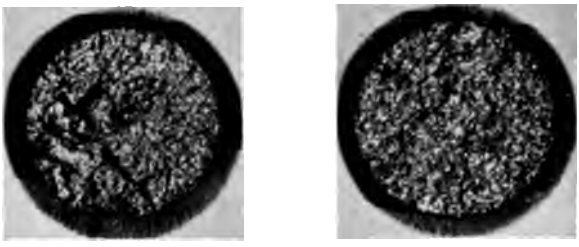


FIG. 21.—TEST-BAR FRACTURES, MUZZLE AND BREECH OF FORGING 1348 UNTREATED. SEE FIGS. 36 TO 41. × 2.

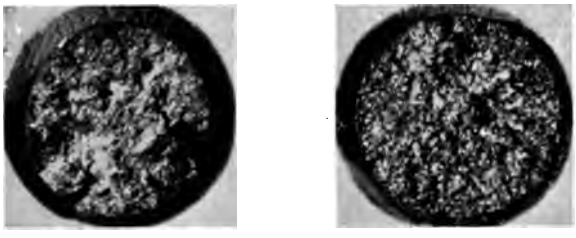


FIG. 22.—TEST-BAR FRACTURES, MUZZLE AND BREECH OF FORGING 1351 UNTREATED. SEE FIGS. 30 TO 35. × 2.

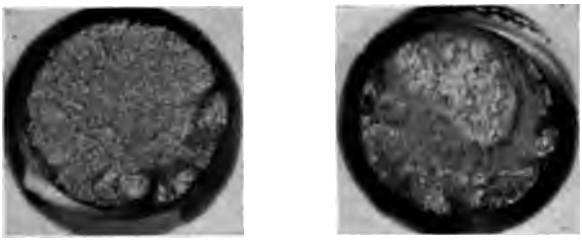


FIG. 23.—TEST-BAR FRACTURES FROM TREATED FORGING 948C. × 2.

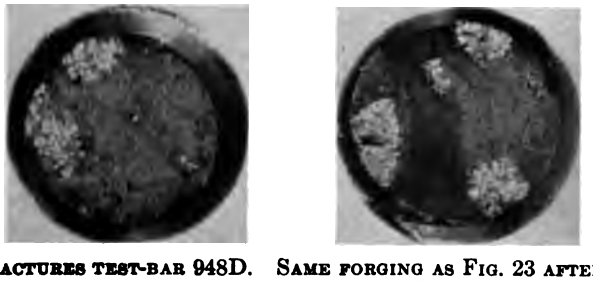


FIG. 24.—FRACTURES TEST-BAR 948D. SAME FORGING AS FIG. 23 AFTER ADDITIONAL QUENCH AND DRAW. × 2.

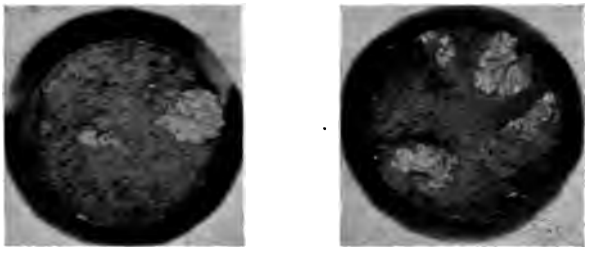


FIG. 25.—FRACTURES TEST-BAR 948B. SAME FORGING AS FIG. 23 AFTER TWO QUENCHES AND A DRAW. × 2.

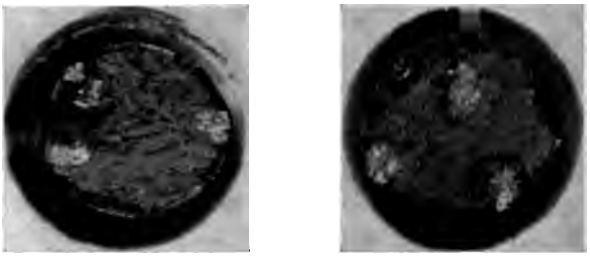


FIG. 26.—FRACTURES TEST-BAR 948A. SAME FORGING AS FIG. 23 AFTER AN ANNEAL TWO QUENCHES AND A DRAW. × 2.

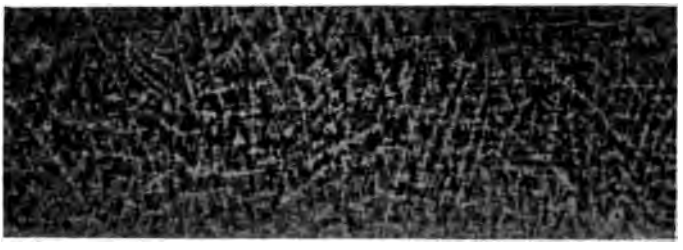


FIG. 27.—DENDRITIC STRUCTURE. SPEC. 948C FROM FORGING THAT HAD BEEN ANNEALED, QUENCHED AND DRAWN. ETCHED WITH STEAD REAGENT. × 2½.

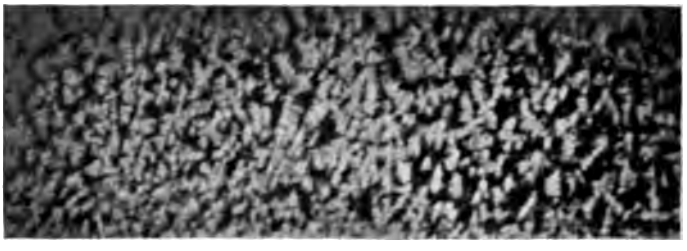


FIG. 28.—DENDRITIC STRUCTURE. SPEC. 948A FROM SAME FORGING AS FIG. 23, AFTER AN ANNEAL, TWO QUENCHES AND A DRAW. ETCHED WITH STEAD REAGENT. × 2½.

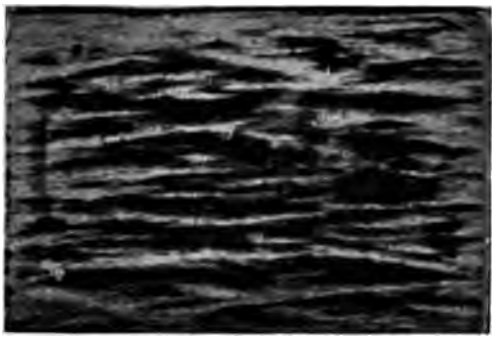


FIG. 29.—DENDRITIC STRUCTURE. EFFECT OF HOT-WORKING ON DENDRITES. SPECIMEN FROM ACCEPTED HEAT-TREATED GUN FORGING. ETCHED WITH STEAD REAGENT. × 2⅝.

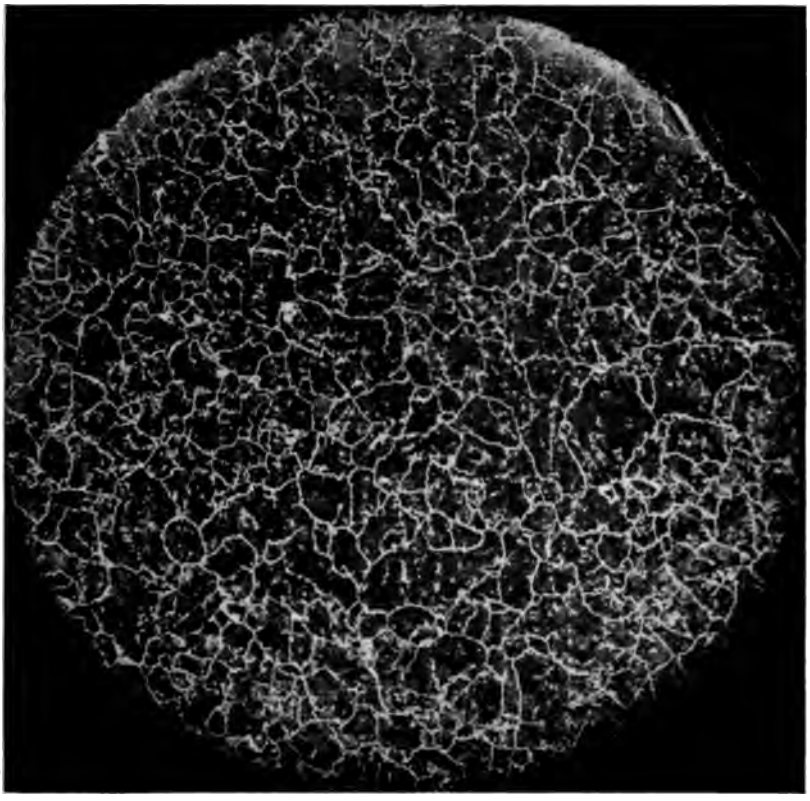


FIG. 30.—CROSS-SECTION OF SCREW END OF TEST BAR FROM FORGING 1351, TAKEN OUT OF BREECH AND OF FORGING, UNTREATED. PICRIC ACID ETCH. × 6. PHYSICAL PROPERTIES: T.S. 90500; EL. 55000; EXT. 5.0 PER CENT.; CONT. 3.9 PER CENT.

Flakes and Flaky Fracture

It is believed that the terms flake and snowflake were first used by ordnance inspectors to designate the peculiar bright crystalline to granular silvery spots found in the fractured ends of pulled test bars from cannon forgings containing the type of defect now under consideration. The term flake has been rather loosely used and has been applied to any bright spots or areas in the fracture of broken test pieces. While possibly not the most desirable name for the defect, it is at least some-

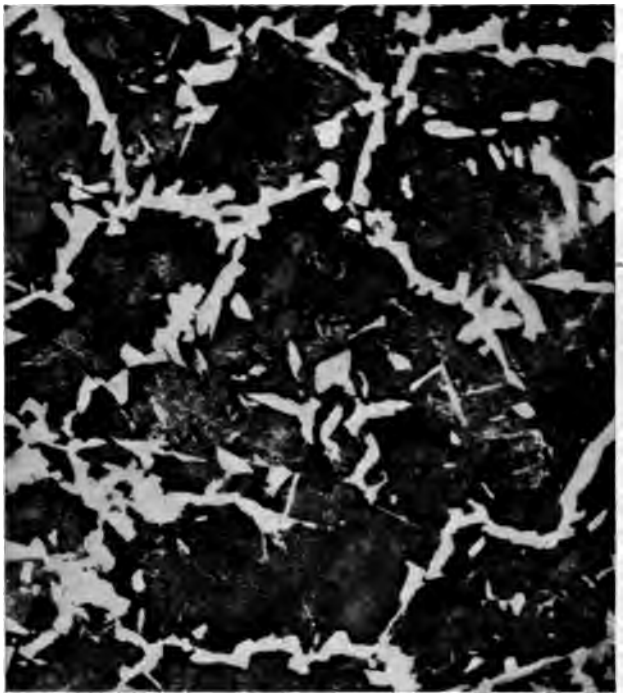


Fig. 31.—Detail of Fig. 30. Picric acid etch. × 90.

what descriptive of the phenomenon, is widely used, and appears to have become a fixture in metallographic nomenclature. The writers are therefore accepting it but are attempting to describe the defect to which it has most consistently been applied as clearly as possible so that it will be limited to this one type of defect.

To the unaided eye, a flaky fracture is any type of fracture in which occur bright, silvery spots or areas. These spots vary usually from $\frac{1}{16}$ to $\frac{1}{2}$ in. (1.5 to 12.7 mm.) in longest diameter and are surrounded by the usual gray to smoke-colored metal, which may be woody, silky, or granular, the silvery spots being the flakes, or snowflakes. A closer examination with a hand lens, or better with a binocular microscope,

shows that the flake appears to be made up of more coarsely granular material than the surrounding metal and in many instances what appear to be the boundaries of polyhedral grains are distinctly visible. These are in the midst of metal of similar color or sheen but without definite crystalline structure. Much of nearly every flake examined in detail was made up of this apparently non-crystalline, or massive, metal in which the fracture presented an irregular nondescript surface. In the definitely crystalline parts of flakes, it was clear that the fracture had

FIG. 32.—DETAIL OF FIG. 30. PICRIC ACID ETCH. × 90.

occurred in two ways, intercrystalline and transcrystalline. That is, the path of the rupture was partly in the ferrite between the grains or areas of sorbite and partly through them.

All flaky steel examined microscopically showed much unmistakable evidence of overheating, and the areas or grains of sorbite were characterized by a prominent development of the Widmannstättian structure. It is probable that in transcrystalline fracture the path of rupture followed the plates of ferrite that produce the Widmannstättian structure in the grains of sorbite. Photographs of fractured ends of test bars showing typical flakes are illustrated in Figs. 7, 11, 23, 24, 25, and 26.

Flakes may be detected also by any of the usual acid etches for macro-

scopic examination of the metal. The flake areas are more readily attacked by the acid and they consequently etch out, leaving tiny open spaces resembling minute cracks. They are so thin that in order to develop them to the best advantage the metal must be surfaced very carefully before etching. This may be explained also on the basis that the acid does not attack the flake itself but rather the metal surrounding it. From this fact it would appear that the flake differs in composition and texture from the surrounding metal.

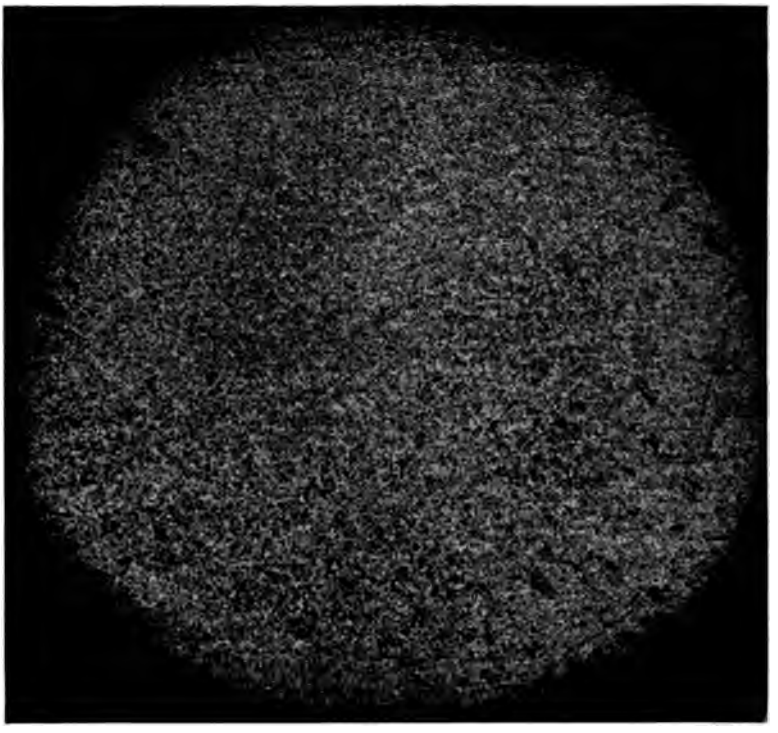


FIG. 33.—CROSS-SECTION OF SCREW END OF TEST BAR FROM FORGING 1351 FROM MUZZLE END OF FORGING UNTREATED. PICRIC ACID ETCH. × 6. PHYSICAL PROPERTIES: T.S. 99000; E.L. 56000; EXT. 18 PER CENT.; CONT. 20.5 PER CENT.

The same areas may be detected also by etching a carefully polished surface with any of the copper-bearing solutions. This differential attack by etch mediums, together with the fact that the flakes in a fracture do not oxidize nearly so readily as the surrounding metal, indicates that the flake differs in composition from the surrounding normal metal. This phase of the subject is discussed in the paragraph on the chemical composition of flakes.

Microstructure

The most characteristic feature of flaky steel revealed under the microscope is the lack of complete grain refinement, see Figs. 8, 12, and 16. Coupled with this and, so far as the writers' experience has extended, invariably present is unmistakable evidence of overheating, such as Widmannstättian structure in the sorbite grains and the so-called spiny ferrite bands. These features are illustrated in Figs. 8, 12, 16, 48, 49, and 50.

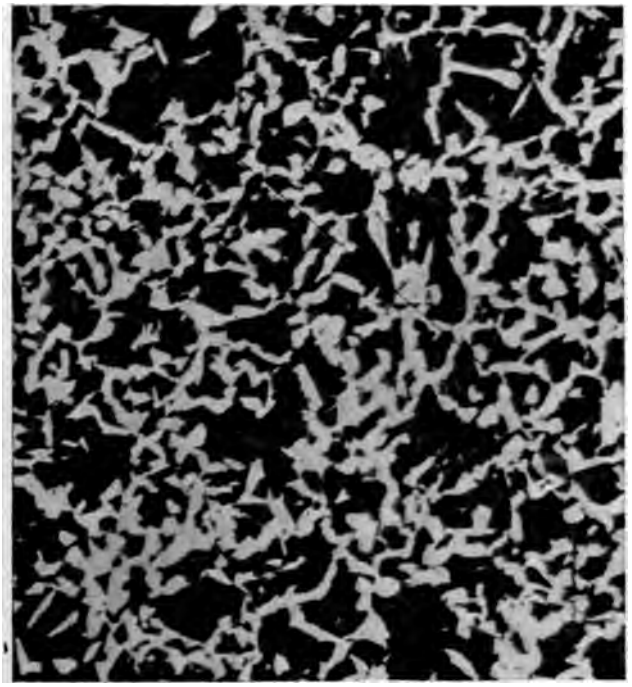


Fig. 34.—Detail of Fig. 33. Picric acid etch. × 90.

Bearing in mind the marked contrast between the flake and the normal metal in a broken test piece, one would expect the microscope to reveal segregations or other physical peculiarities. But only in a few instances were any such segregations found and in still fewer cases were these definitely related to flakes, and even then the relation was not such as would warrant the conclusion that the flake, as revealed in the fracture, and the segregation were identical. The microscope, in addition to showing the lack of grain refinement and the prominence of Widmannstättian structure in flaky steel, also reveals certian peculiar areas generally occurring as irregular connected zones or patches in which the grains of sorbite and the areas of ferrite are different from those of the surround-

ing metal. In all such instances examined the grains appear to be smaller and with possibly a little heavier precipitation of ferrite, as well as more prominent Widmannstättian structure, than the other part of the metal. These features are very difficult to describe but they are easily discernible under the microscope and are illustrated in Figs. 42 to 45. While microscopic examination does not provide an answer as to the physical nature of the flakes, it does indicate that they are caused by the rupture of small areas irregularly distributed through the metal so differing from

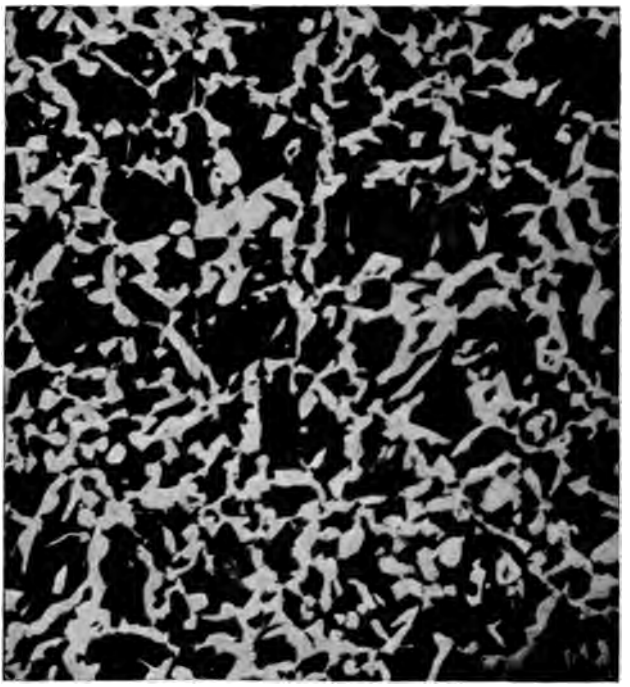


FIG. 35.—DETAIL OF FIG. 33. PICRIC ACID ETCH. × 90.

the other metal that the rupture occurs both in the ferrite and through the sorbite along the Widmannstättian plates. It might be surmised that these areas represent places in which some non-ferrous element has segregated in the ferrite, thus making it more brittle than the surrounding metal, such, for example, as phosphorus, also lowering the temperature at which these particular areas would be overheated.

CHEMICAL TESTS

It has not been possible to isolate enough metal from the flakes for quantitative or even satisfactory qualitative analysis; the writers hope to do this in the near future. However, a certain amount of fairly

reliable qualitative and microchemical work has been done, all of which indicates that the flake areas are higher in phosphorus and carbon than the normal metal. In etching flaky steel for macroscopic examination, it was noted, as is illustrated in Figs. 46 and 47, that the flake areas were attacked more readily than the normal steel and if the sample was permitted to remain long in the acid the flake area was eaten out and resembled a minute crack. Also, when a carefully polished surface of flaky steel was etched with any of the copper-bearing etch mediums, the results invariably indicated that the flake areas differed in composition or condition from the rest of the metal.

Effect of Flakes

The physical properties determined upon the bars coming to the writers' attention are elastic limit, tensile strength, elongation, and reduction in area. In applying the load on a tensile machine, an interesting phenomenon was observed. As shown in Fig. 19, throughout the bar small cracks developed at right angles to the direction of stress; these cracks are without doubt flake areas. It is perhaps this phenomenon that has led some investigators to conclude that flakes are incipient cracks.

The most characteristic effects of flakes, as found in these bars, is the loss of ductility and corresponding decrease in reduction of area; the tensile strength and elastic limit are reduced but not so marked as these other properties. This is best shown by reference to bars 36, 38, and 032, which gave an elongation of 3.6, 6.8, and 0.0 per cent. and a reduction in area of 2.0, 4.5 and 0.0 per cent. (see Table 1). The corresponding tensile strengths are 62,900, 108,100 and 47,000 lb. (28,531, 49,033 and 21,318 kg.), respectively. The decrease in physical properties may or may not be dependent on the size of flakes. Generally speaking, the

Table 1.—*Chemical and Physical Properties of Test Bars*

No.	Type of Fracture	Chemical Analyses						Physical Properties			
		C, Per Cent.	Mn, Per Cent.	Si, Per Cent.	S, Per Cent.	P, Per Cent.	Ni, Per Cent.	Elastic Limit, Pounds Per Sq. In.	Tensile Strength, Pounds Per Sq. In.	Elongation, Per Cent.	Reduction in Area, Per Cent.
16	Cup	0.39	0.66	0.25	0.025	0.020	2.90	73,900	97,200	29.5	64.7
7	Star	0.41	0.79	0.29	0.011	0.020	2.87	78,700	108,600	24.5	63.5
8	Woody	0.42	0.60	0.122	0.019	0.019	3.14	72,800	87,000	5.5	17.7
18	Woody	0.41	0.66	0.25	0.015	0.020	2.98	66,700	115,900	19.5	39.5
20	Woody	0.40	0.55	0.15	0.021	0.025	2.85	62,900	97,300	7.0	7.8
36	Flake	0.44	0.66	0.34	0.021	0.020	3.29	61,900	62,900	2.0	3.6
38	Flake	0.37	0.59	0.19	0.019	0.020	3.09	73,400	108,100	4.5	6.8
032	Flake	0.38	0.69	0.24	0.033	0.043	3.24	47,000	47,000	0.0	0.0
39	Slag	0.41	0.66	0.23	0.040	0.036	3.27	77,000	106,600	9.0	11.8

smaller the flakes the better the steel but this is not always true, nor can it be said that these properties and the size of flakes are always definitely related. This is best shown by comparing bars 032 and 36 (Figs. 11 and 7). These bars show practically the same area of flake in the fracture, about 50 per cent., yet bar 032 shows a tensile strength of only 47,000 lb. while bar 36 shows a strength of 62,900 lb. The nature of the surface of the flake, that is, whether fine or coarse grained, may affect the physical properties and may explain the difference in properties in bar 032, which is very coarse grained, and bar 36, which is finer grained.

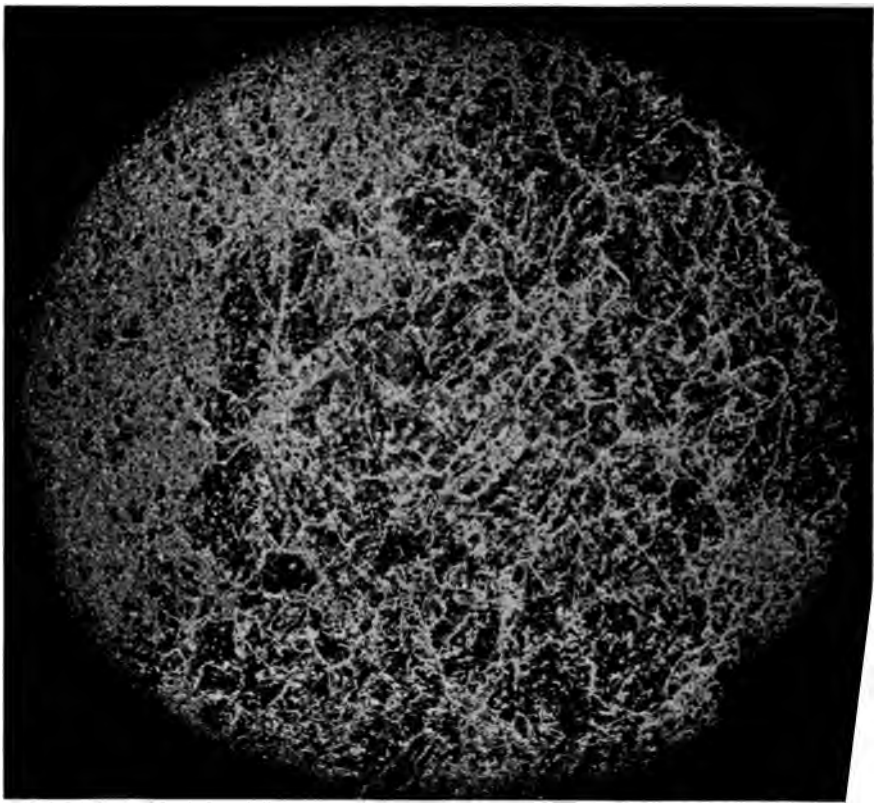


FIG. 36.—CROSS-SECTION OF SCREW AND OF TEST BAR FROM FORGING 1348 FR MUZZLE END OF FORGING UNTREATED. PICRIC ACID ETCH. × 6. PHYSICAL PROP TIES: T.S. 92000; E.L. 55000; EXT. 11 PER CENT.; CONT. 13.3 PER CENT.

SUGGESTED CAUSES OF FLAKES

The following suggestions have come to the attention of the wri as possible causes of flakes:

1. Foreign inclusions. Microexamination of a number of defe bars has failed to tie up in a conclusive manner foreign inclusions

flaking. In fact, many such bars have been found that show the steel to be exceptionally clean in this respect.

2. Gaseous inclusions and cavities. Little investigation has been done along this line, but here again there has been little microscopic evidence that would warrant extensive research in this direction.

3. Scale and consequent local decarburization. It has been suggested that in rapid pouring the oxide film that forms on top of the metal may be churned in and entrapped in the metal on freezing. One would expect

FIG. 37.—DETAIL OF FIG. 36. PICRIC ACID ETCH. × 90.

in this event to find oxide inclusions but the writers have not found this to be so in even a majority of cases, and while in some instances oxides have been found they have not appeared in sufficient quantity to set this down as a primary cause of flaking.

4. Segregation of non-ferrous elements. The writers' experience in this matter is covered in the part of this paper dealing with the chemical composition of flakes.

5. Overheating and non-uniformity of heat of ingot preparatory to forging. This seems to the writers to be on fairly firm ground and is discussed at greater length elsewhere in this paper.

6. Shape of ingot and method of forging. In many cases square ingots forged on a flat die have been used. A change to octagon-shaped molds and forging in a V-shaped die decreased rejections to some extent; but as far as the writers' information goes, this change has not resulted in completely doing away with flaking.

7. Differential expansion of areas in heterogeneous metal. The suggestion is made that, where there are areas with varying coefficients of expansion, due to the more rapid expansion and contraction of certain

FIG. 38.—DETAIL OF FIG. 36. PICRIC ACID ETCH. × 90.

parts in heating and cooling, lines of weakness or even ruptures may be set up in the mass which upon being stressed break readily with low ductility.

CAUSE OF WOODY FRACTURE

Practically all suggestions as to the cause of woody fracture emanate from the same line of thought, that they are caused by the condition of the steel with respect to the original dendritic structure (Fig. 27) of the ingot; this appears quite a plausible theory. The theory has been advanced that slow cooling through the solidification zone will lead to large dendrites that, upon being drawn out at the forge, form long plates.

Rapid cooling through the solidification zone has been suggested as a remedy but the rapidity of the cooling must be governed by the chances of setting up bad secondary shrinkage in the ingot.

It is probable that laminated or woody fiber is produced, in a great measure, by the working of the original dendrites (Fig. 29) in parallel longitudinal lines by forging. The greater the amount of work applied,

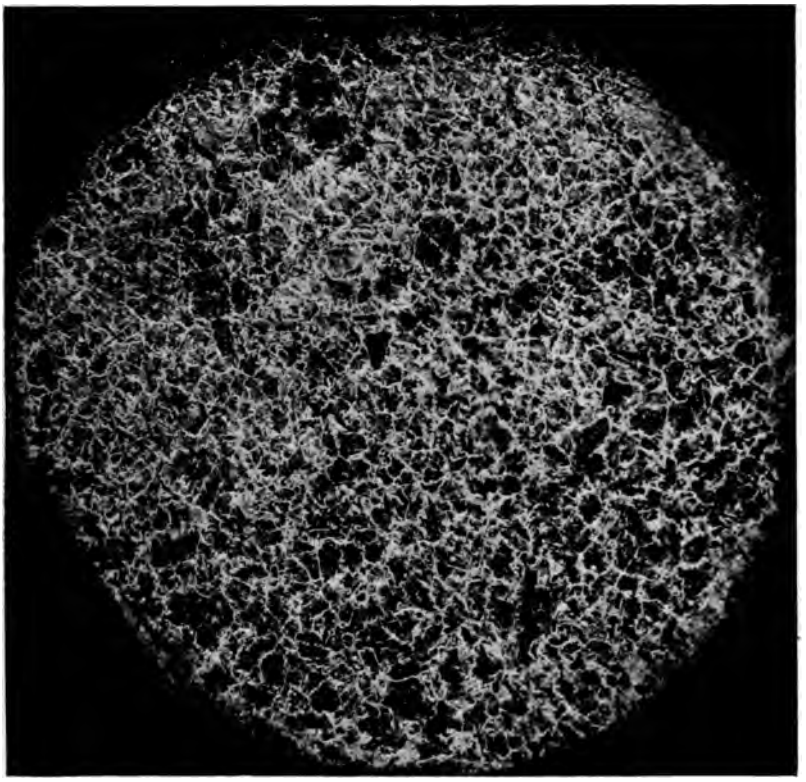


FIG. 39.—CROSS-SECTION OF SCREW END OF TEST BAR FROM FORGING 1348 FROM BREECH END OF FORGING UNTREATED. PICRIC ACID ETCH. × 6. PHYSICAL PROPERTIES: T.S. 90000; E.L. 50000; EXT. 8 PER CENT.; CONT. 9.6 PER CENT.

the longer and thinner these lines become; here may be found the cause of the exceptionally low transverse ductility in bars that show a decidedly laminated or woody structure in which fibers can be seen very plainly in the macrostructure (Fig. 29). Charpy[1] has shown the bad effect of hot working on the transverse ductility of steel and Portevin and Bernard[2]

[1] G. Charpy: Influence of Hot Deformation on the Qualities of Steel. *Engng.* [London] (Sept. 20, 1918) **106**, 310.

[2] A. Portevin et V. Bernard: La Macrostructure de l'Acier. *Rev. de Mét.* (1918) **15**, 273–280.

were able to efface the remains of original dendritic structure only by annealing for 6 days at 850° C. (1562° F.) to 900° C. (1652° F.).

The persistence of dendritic structure after fairly severe treatment may be seen in Fig. 28. In this connection it may be said that anything which leads toward the production of a coarse fiber is undesirable. A basic steel, generally speaking, will fiber more readily than an acid steel and for this reason basic steel is desirable for armor where a fibrous back is essential. In the manufacture of nickel-steel gun forgings, where a coarse fiber is not desirable, acid steel has been most successfully used.

FIG. 40.—DETAIL OF FIG. 39. PICRIC ACID ETCH. × 90.

Hot working will deform dendrites and tend to form them into parallel bands, therefore, where transverse strains are to be withstood, hot working should be carried on only to the point that experience will dictate. This would suggest the use of longer ingots of smaller cross-section. But with ordinary plant practice it does not appear certain that such a change would not be jumping from the frying pan into the fire. Forging seems to be carried out, in a great many instances, at excessive temperature, resulting in finishing at too high a heat, so that with less work, due to the reduced size of the ingot, the dangers arising from high finishing

temperatures would be even greater and result in worse condition than could be brought about by an excessive deformation of dendrites.

Figs. 30 to 32 and 36 to 41 show the effect of overheating and high finishing temperature. Figs. 33 to 35 show what may be expected from a lower forging temperature and consequently lower finishing temperature. The absence of Widmannstättian structure in the high-power photomicrographs (Figs. 34 and 35) is noteworthy as compared with Figs. 31, 32, 37, 38, 40, and 41. This evidence of overheating is apparent

FIG. 41.—DETAIL OF FIG. 39. PICRIC ACID ETCH. × 90.

even after fairly prolonged heat treatment and is prevalent in the sorbite of bars that show flakes. Forging can be done at much lower temperatures and, where the reduction at the hammer is slight, there is no reason why the forging temperature cannot be kept as low as 1038° C. (1900° F.). It is possible that the defects which have caused a great number of forgings to fail to meet specifications might be traced to overheating in forging, temperatures in excess of 1371° C. (2500° F.) having been measured in furnaces in which ingots had been for as long as 8 hr. In this particular case, the bottom end of the ingot was in the hottest zone and records of the plant show 70 per cent. defective bars from this end of the forging compared with 46 per cent. in the other end of the forging, which

came from about half way up in the ingot. That these defects result from incipient fusion is a reasonable supposition. By reference to the carbon-iron diagram, the solidus at 0.40° C. (the steel in question was about 0.38 per cent. carbon and 3 per cent. nickel) is found to be not greater than 1340° C. (2444° F.), or 30° C. lower than the temperature observed above. Little has been done with the iron-nickel equilibrium diagram, but it is understood that nickel lowers the solidus line; therefore, more care must be taken with respect to the forging temperature of

FIG. 42.—FOR DESCRIPTION, SEE FIG. 45.

nickel steel, and successful forging of carbon steel does not imply that good forgings will follow the same practice with nickel steel. Once the steel has been heated above its solidus, small spots of fused metal will be formed throughout the steel where the concentration of carbon is greatest. At these points the original dendrites are at least partly destroyed and, upon forging, are drawn out between the adjacent dendrites, where they exist as burnt steel, forming lines of weakness and low ductility. That the damage which causes flakes is done after the ingot has been poured is indicated strongly by an experiment in which two ingots were cast from the bottom through a common runner; one ingot made a flaky forging and there was no trouble from this source in the other.

It must be remembered that, in cooling from the molten state through the Ar transformations, a very large grain size is set up with thick ferrite envelopes. No one would expect to obtain physical properties nor any fair degree of grain refinement to result from the heat-treating of ingot steel as cast. The structure does not respond, owing to the difficulty of getting the ferrite into solution. Very likely, in this case, the austenite is quite heterogeneous due to the slow diffusion of the ferrite from the thick boundaries of the original grains into the austenite grains formed in the pearlite, resulting in the grains that formed close to the ferrite

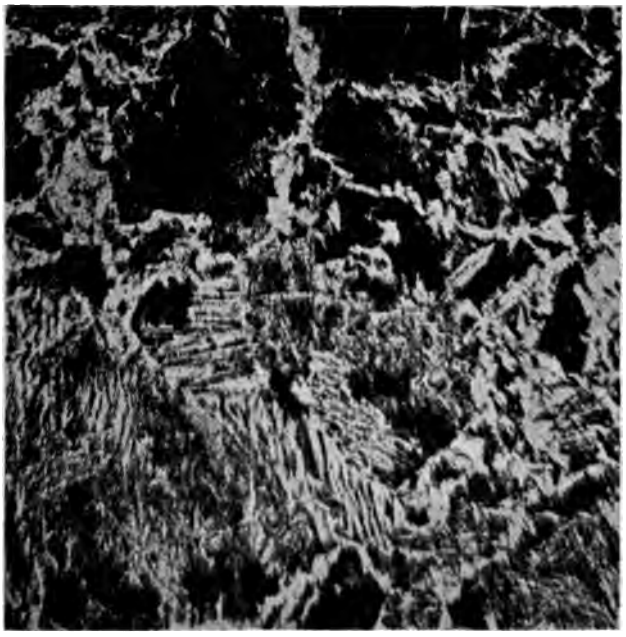


FIG. 43.—FOR DESCRIPTION, SEE FIG. 45.

envelope of the original grain being higher in iron content. This is the condition of the austenite that may exist upon reheating ingots for forging. In the case of a steel containing 0.38 per cent. carbon heating from below Ar, the first transformation will be that of the pearlite to austenite at Ac_1, followed by the absorption of the α iron envelope until at $Ac_{2,3}$ the envelope has disappeared. It does not necessarily follow, however, that austenite immediately below $Ac_{2,3}$ with its dissolved α iron is homogeneous. In reheating steel of ingot structure, it is likely that the pro-eutectoid iron has not been able to penetrate to the grains of austenite formed at the interior of each of the grains that existed below Ac_1, so that there is still an envelope of iron-rich metal; nor will this heterogeneity end at $Ac_{2,3}$ with the transformation of the α iron into γ

when it enters the solid solution. Diffusion goes on in the austenite, but probably becomes complete only after a considerable length of time and, if the heating is carried to the solidus, there may still exist at that temperature spots of high carbon concentration in the center of what were the old grains of pearlite, these spots now existing as high-carbon austenite grains. These are the grains in which fusion will start and they will very likely be the austenite existing in the center of what were the largest of the original pearlite grains. As the temperature rises above the solidus, here and there throughout the steel there will be molten spots, their size

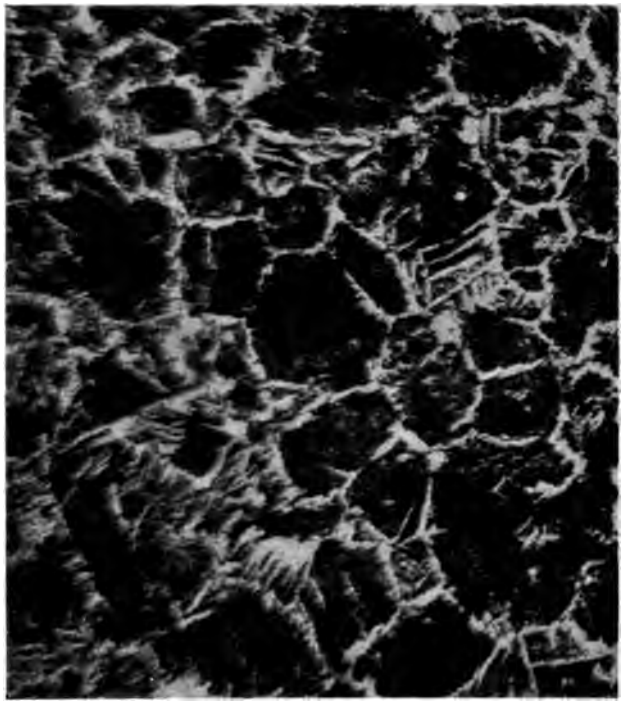


FIG. 44.—FOR DESCRIPTION, SEE FIG. 45.

depending on the temperature attained, surrounded by badly overheated metal, which upon forging will be drawn out flat. The forging may reclaim considerable of the overheated metal but the fused part will persist as films, having been the most plastic during the forging. Upon sorbitizing a steel in this condition, the part that has not been molten during the forging will have responded to treatment, whereas the films where the molten globules have been forged out have been only slightly affected, if at all. In a transverse test these two conditions will be apparent in a ductile, gray, fibrous fracture in which will appear a bright silvery area of low ductility, known as a flake. This bright area may consist of metal

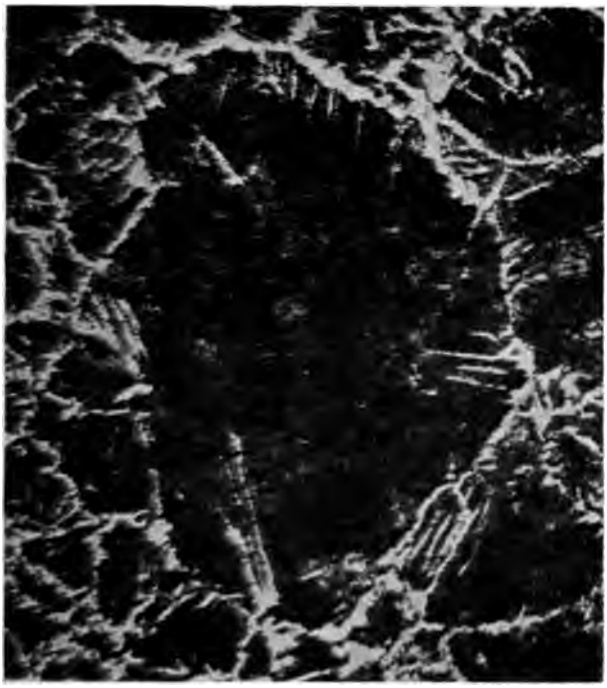


FIG. 45.

FIGS. 42 TO 45.—COARSE NETWORK OF NEEDLE-LIKE FERRITE AND SORBITE. TYPICAL STRUCTURE OF SOME FORGINGS AFTER IMPROPER HEAT TREATMENT. LACK OF UNIFORMITY IN STRUCTURE AND WIDMANNSTÄTTIAN STRUCTURE ARE NOTEWORTHY.

FIG. 46.—SURFACE OF FLAKY STEEL AFTER MACROSCOPIC ETCHING AND SLIGHT REPOLISHING. APPARENT CRACKS ASSUMED TO BE FLAKE AREAS EATEN OUT BY THE ACID. × 20

in various conditions, ranging from burnt to only slight overheating. Severe treatment may reduce the size of the bright areas in so far as it is able to reclaim that part which has been overheated, but there will

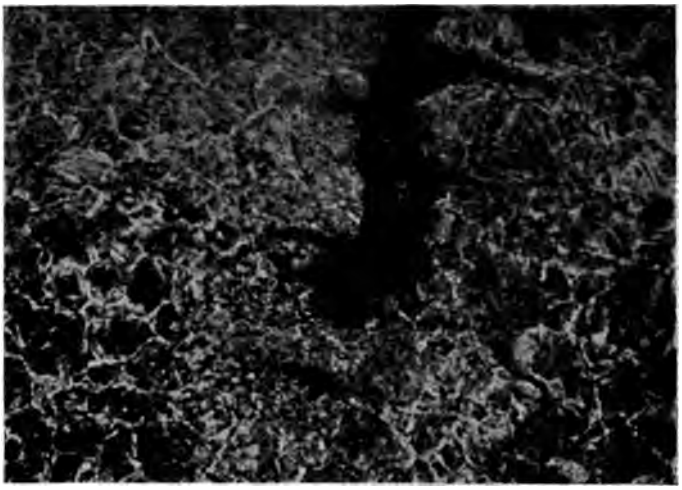


FIG. 47.—PORTION OF AREA SHOWN IN FIG. 46 AFTER REGRINDING AND ETCHING WITH PICRIC ACID. × 100.

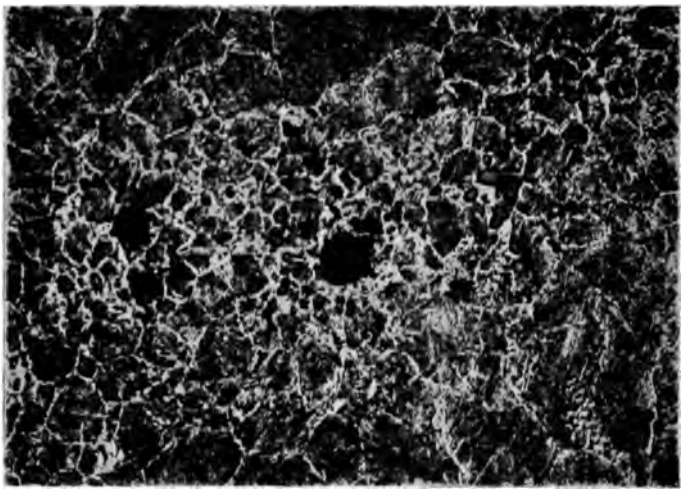


FIG. 48.—SPECIMEN OF FIGS. 46 AND 47 AFTER DEEPER GRINDING. PICRIC ACID ETCH. × 100.

always remain a part that has been fused, and this may never respond. The burning that produced this condition need not, and probably rarely does, become evident on the surface of the ingot for the reason that, in most cases, heating is done in an oxidizing atmosphere and the surface

of the ingot is decarburized, thus giving a coating of high melting point metal to the ingot.

It will be understood, from the foregoing hypothesis, that, in case of a

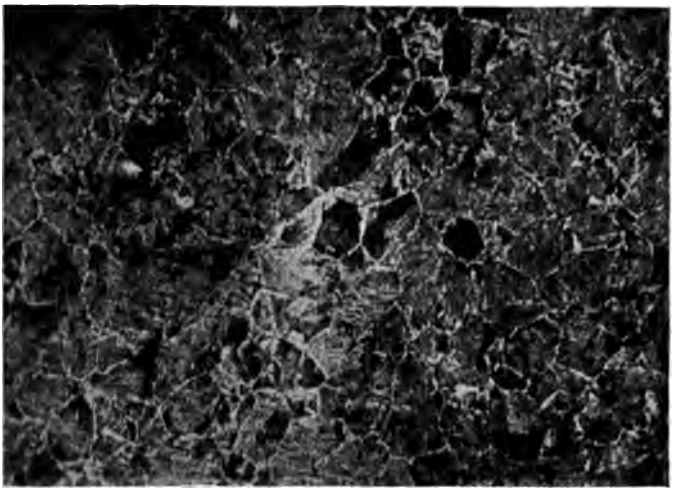


FIG. 49.—ANOTHER SPECIMEN OF SIMILAR METAL AFTER ETCHING WITH PICRIC ACID SHOWS AREA WHICH MANIFESTED ITSELF IN THE FRACTURE AS A FLAKE. × 100.

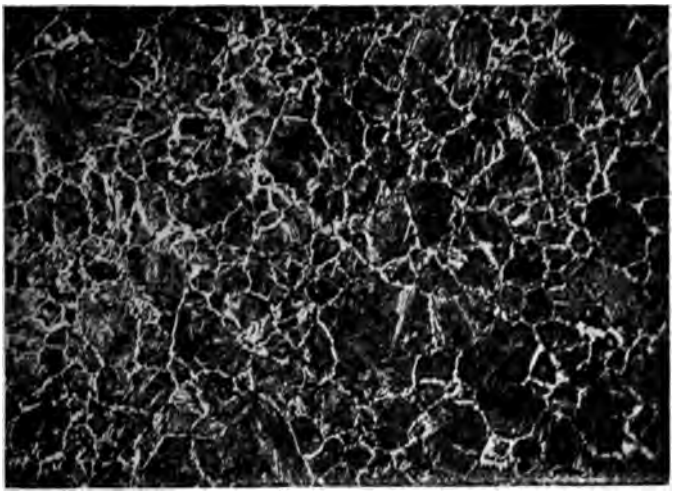


FIG. 50.—SPECIMEN OF METAL SIMILAR TO THAT IN FIGS. 46 TO 49, SHOWING MICROSTRUCTURE ASSOCIATED WITH FLAKY FRACTURE, ETCHED WITH PICRIC ACID. × 100.

heterogeneous austenite, one is not so much concerned with the solidus of the steel as a whole but with the solidus corresponding to that of the austenite of highest carbon content, which, in many cases, may be of

eutectoid composition. To get away completely from any possibility of this condition, forgings should not be heated to a temperature exceeding 1220° C. (2228° F.) and if a higher temperature is necessary great care should be exercised to see that the ingot is brought to the temperature slowly and well soaked out as the temperature rises, in order to permit of as much diffusion as possible.

The following attempt was made to efface flakes from material known to be very bad in this respect. Disks 15 in. (38 cm.) in diameter by 1¼ in. (3.18 cm.) thick with a 5⅜ in. (13.7 cm.) bore were obtained from each of two forgings that had been condemned for excessive flakiness. Each disk was cut along two diameters, making in all eight quadrants. These pieces were marked 948-A, B, C, and D and 986-A, B, C, and D, and all, except 948-C and 986-D and C, were treated. After treatment middle and inside test bars were machined out according to Ordnance Department specifications. The results are given in Table 2.

TABLE 2.—*Results of Tests of Condemned Forgings*

Piece No.	Treatment			Physical Properties				Position of Bar Tested
	Heated to Degrees C.	Held at Temperature, Hours	Cooled in	Tensile Strength, Pounds	Elastic Limit, Pounds	Elongation, Per Cent.	Reduction in Area, Per Cent.	
948–C	Untreated			80,530	54,770	3.0	10.0	Inside
				88,330	48,100	4.5	8.5	Middle
948–D	760	2	Water	93,140	71,610	6.5	17.5	Inside
	649	2	Furnace	93,760	78,570	4.5	16.5	Middle
948–B	815	2	Water					
	760	2	Water	96,220	65,560	6.5	13.0	Inside
	649	2	Furnace	94,870	65,430	7.0	14.0	Middle
948–A	871	2	Furnace					
	815	2	Water					
	760	2	Water	102,300	81,000	12.5	27.0	Inside
	649	2	Furnace	101,550	73,930	10.5	19.0	Middle
986–D	Untreated			88,000	53,500	5.0	10.0	Inside
				77,000	51,000	5.0	3.0	Middle
986–A	954	6	Air					
	760	4	Water	85,640	54,140	8.0	20.5	Inside
	649	3	Furnace	96,500	60,000	12.5	22.0	Middle
986–B	1024	2	Furnace					
	871	2	Furnace					
	760	2	Water	96,500	78,500	7.5	18.1	Inside
	649	2	Furnace	90,500	78,500	5.0	13.4	Middle

Analyses showed that the disks marked 948 contained 0.40 per cent. carbon, 0.62 per cent. manganese, 0.018 per cent. phosphorus, 0.027 per cent. sulfur, 0.12 per cent. silicon, 2.99 per cent. nickel, and 0.14 per cent. chromium. The disks marked 986 contained 0.34 per cent. carbon, 0.55 per cent. manganese, 0.028 per cent. phosphorus, 0.019 per cent. sulfur, 0.10 per cent. silicon, 2.88 per cent. nickel, and 0.16 per cent. chromium. Pieces 948 showed a decided improvement in both

physical properties and flaking, the three heatings through the critical temperature giving the best results. Fig. 23 shows the condition as to flakes of disk 948–C; Fig. 24, the condition of disk 948–D; Fig. 25, the condition of disk 948–B; and Fig. 26, the condition of disk 948–A. The microstructure of 948–A is shown in Fig. 20; the dendritic structure in 948–C is shown in Fig. 27, and its persistence after treatment, in Fig. 28. There was no indication that heating to a higher temperature and holding for a long time at that temperature was of great benefit; this is apparent from results obtained in the case of pieces 986. The specifications called for a tensile strength of 95,000 lb., an elastic limit of 65,000 lb., an elongation of 18 per cent., and a reduction of area of 30 per cent.

Discussion of this paper begins on page 263.

Microstructural Features of Flaky Steel*

BY HENRY S. RAWDON,† WASHINGTON, D. C.

(New York Meeting, February, 1919)

ONE of the most vital problems in the manufacture of steel at present is the occurrence of the defects that have been popularly termed "snow flakes," "flakes," or "scabs." Particularly is this the problem of many manufacturers who, during the past 2 yr., have undertaken large-scale production for the first time. A study of the microstructure of such defective material throws considerable light upon the nature of these defects and is essential before measures may be taken for the elimination of the defects. Flakes do not appear to be found in the plain carbon steels; they have been found, however, in abundance in some of the simple alloy steels that are being used in such vast quantities; viz., nickel steel of the approximate composition, 0.40 carbon and 3.5 per cent. nickel, and in the chrome-nickel type. The present article summarizes the characteristic features of defective steel of the flaky type as they have been found by laboratory study of numerous specimens, and it aims to show the conditions within the metal that are favorable to the occurrence of this type of defect without entering into the details of mill practice (size of ingot, design of molds, rate of cooling, distribution of cooling stresses, etc.), all of which conditions appear to play a role in the production of these defects.

MACROSTRUCTURE

To one familar with flaky steel, the appearance of "snow-flakes" is unmistakable and is not to be confused with other types of defects. Figs. 1 to 6 show the appearance of fractures of steel showing this type of defect; the white silvery area, which always has the appearance of being of a very coarsely crystalline structure, in the specimen stands out in bold contrast to the darker background and readily justifies the use of the term "snow-flakes." Fig. 1 shows the appearance of the faces of the two halves of a nickel-steel forging that has been split to show the interior flakes. Fig. 2 is a nickel-steel bar (carbon 0.40 per cent.,

* Report of research under the Auspices of the National Research Council.

† Associate Physicist (Metallography), U. S. Bureau of Standards.

nickel (3.62 per cent., chromium 0.35 per cent. not intentionally added) intended for use in the manufacture of a propeller shaft.

FIG. 1.—NICKEL-STEEL FORGING SPLIT TO REVEAL FLAKES. × 1.

In the examination of material for the detection of flakes, the specimens are often nicked and broken under impact; unless the steel has a very

FIG. 2.—CHROME-NICKEL STEEL BAR, C, 0.40; NI, 3.62; CR, 0.35, THAT HAS BEEN NICKED AND BROKEN TO SHOW FLAKES. × 1.

fine grain the contrast between the coarse face of the flake and the surrounding metal may not be very marked. The best procedure is to refine the grain by properly hardening the piece before breaking. The

flakes will then appear as coarsely crystalline areas surrounded by the smooth porcelain-like fracture of the surrounding sound metal. Fig. 3 shows such a fracture of the material of Fig. 2 after so hardening. The

FIG. 3.—MATERIAL SIMILAR TO THAT OF FIG. 2, THAT HAS BEEN HARDENED AND THEN BROKEN. × 1.

area designated as the flake always has an appearance suggesting a coarsely crystalline structure. The crystals, however, are markedly

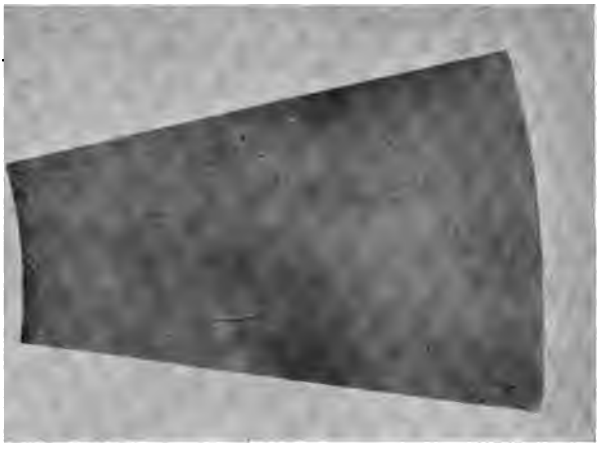


FIG. 4.—SECTION OF DISK CUT FROM NICKEL-STEEL GUN FORGING, C, 0.38, CR, 0.20, NI, 2.92, POLISHED TO SHOW FLAKES. × 1.

different from those of the remainder of the fracture—no sparkling cleavage planes are to be found, instead the corners and edges of the crystals are rounded and smoothed and the surfaces often have a very

fine matt finish. They have the appearance of crystals that were squeezed together while soft and plastic and yet failed to adhere firmly to one another. The same type of break is obtained by fracturing a bar

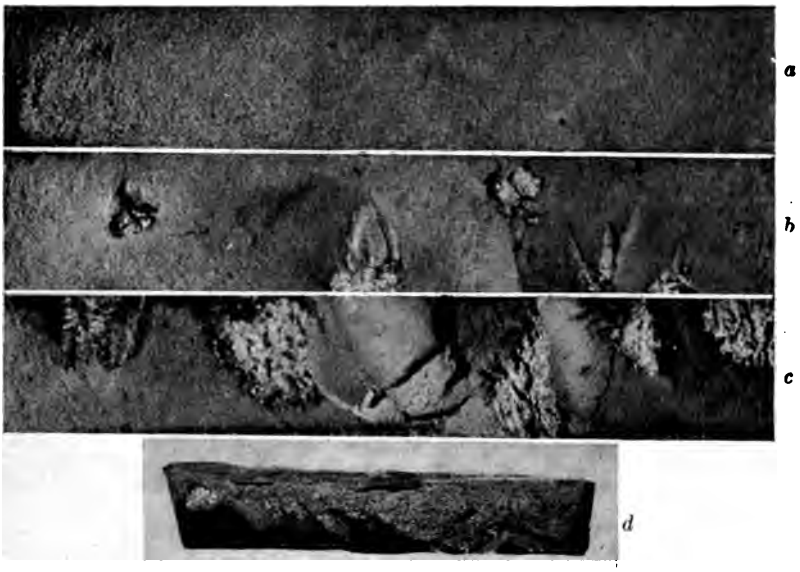


FIG. 5.—*a*, BAR CUT TRANSVERSELY FROM NICKEL-STEEL GUN FORGING, BROKEN BY ALTERNATING STRESSES; MATERIAL WAS SOUND, NO FLAKES WERE REVEALED. × 2. *b*. SIMILAR MATERIAL TO *a*; SEVERAL FLAKES WERE REVEALED WHEN BROKEN BY ALTERNATING STRESSES. × 2. *c*. SIMILAR MATERIAL TO *a* AND *b*; METAL CONTAINS NUMEROUS LARGE FLAKES. SOUND METAL FORMS SMOOTH PORCELAIN FRACTURE, FLAKES HAVE USUAL COARSELY CRYSTALLINE APPEARANCE. × 2. *d*. SPECIMEN SIMILAR TO *b* BROKEN BY IMPACT; FLAKES ARE RELATIVELY INCONSPICUOUS, COMPARED WITH THOSE IN *b* AND *c*. × 1.

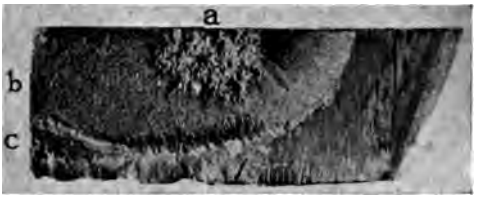


FIG. 6.—MATERIAL OF FIG. 5*c*; THIS FLAKE WAS LOCATED BACK OF PLANE OF FRACTURE AND WAS VERY SLIGHTLY OPENED UP BY ACTION OF ALTERNATING STRESSES; IT WAS THEN BROKEN OPEN BY IMPACT. × 2. *a*, COARSE FRACTURE DUE TO FLAKE. *b*, SMOOTH PORCELAIN DETAIL FRACTURE. *c*, COARSE CRYSTALLINE BREAK DUE TO IMPACT.

by the action of alternating or by repeated stresses. Figs. 5 and 6 show the appearance of a sound specimen of nickel steel, carbon 0.41, chromium 0.106, nickel 2.85, and two flaky bars of similar material,

carbon 0.37, chromium 0.01, nickel 2.83, fractured by alternating stresses. The flakes revealed in this manner are much more striking in appearance than when the same specimen is broken transversely by impact. Fig. 5d shows one of these specimens broken in this latter manner. That the area of the flake is a nucleus for the starting of the

FIG. 7.

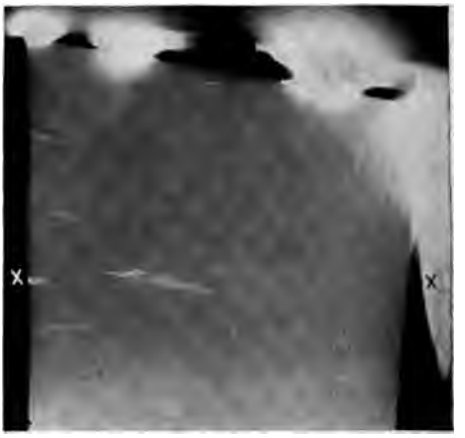


FIG. 8.

FIG. 7.—RADIOGRAPH OF BAR OF SOUND NICKEL-STEEL ⅜ IN. THICK, SAME MATERIAL AS FIG. 5*a*.

FIG. 8.—RADIOGRAPH OF SIMILAR BAR OF FLAKY NICKEL-STEEL, SAME MATERIAL AS FIG. 5*c*; EACH WHITE TRANSVERSE LINE MARKS LOCATION OF A FLAKE.

detail fracture that produces the fine porcelain-like break is illustrated in Fig. 6. This represents a flake located somewhat back of the plane on which the break finally took place in the specimen, Fig. 5c, broken by alternating stresses. After the completion of the test, the metal here was opened by sawing in from the sides and then breaking by a hammer

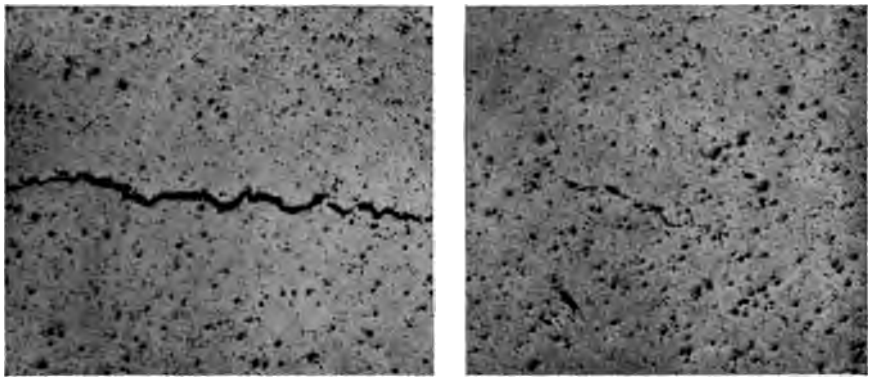
FIG. 9.—CROSS-SECTION OF A BAR OF FIG. 8, BROKEN ALONG THE LINE *x*–*x* IN RADIOGRAPH. × 2.

blow. The fracture shows three portions: the coarse crystalline flake, the smooth detail break surrounding this, and the outer crystalline portion broken by impact. Without doubt this flake acted as a center from which the detail break proceeded outward on all sides.

The flakes are often found to have a rather definite and symmetrical

arrangement in the piece in which they occur. Fig. 4 shows a sector of a cross-section of a gun forging, carbon 0.38, chromium 0.20, nickel 2.92 in which the flakes were very abundant. The radial arrangement is very evident. The same symmetry has been noted in much smaller pieces, that is 4 and 5-in. (10 and 12-cm.) billets of chrome-nickel steel.

The examination by means of X rays of steel suspected of being flaky is of instructive interest and throws some additional light on the nature of the defect. Figs. 7 and 8 show radiographs of a sound $\frac{3}{8}$-in. (9 mm.) bar of steel, carbon 0.41, chromium 0.106, nickel 2.85, and a specimen of very similar composition, carbon 0.37, chromium 0.010, nickel 2.83, but suspected of containing flakes. These bars were cut transversely out of finished gun forgings of the approximate composition



a *b*

FIG. 10.—SAME MATERIAL AS FIG. 4, ETCHED WITH 2 PER CENT. COPPER-AMMONIUM CHLORIDE SOLUTION. × 15. *a*, LARGE FLAKE; WHICH APPEARS TO BEAR NO RELATION TO SLAG INCLUSIONS. *b*, SMALL FLAKE; INCLUSIONS THAT AIDED IN FORMATION ARE PLAINLY SEEN.

given. Each white transverse line in the second radiograph, Fig. 8, indicates the location of a flake as was shown by sawing in from the sides of the line $x - x$ and breaking the specimen along the line. The large flake shown in Fig. 9 was revealed in this manner. The white lines marking the location of the flakes indicate discontinuities in the metal along these lines. They are due to the metal being more transparent to the rays along these planes than is the surrounding metal.

In the examination of flakes to show their relation to the structural features of the steel, it is essential that the study be made upon samples showing very small flakes. Figs. 4 and 8 show that the flake is a discontinuity within the metal often of very appreciable width. Fig. 10a shows a section through one of the flakes in the gun forging shown in Fig. 4. The metal has been etched with copper-ammonium-chloride solution to show the non-metallic or slag inclusions. Although the steel is very

badly contaminated with such inclusions, little if any relation between these inclusions and the course of the flake can be traced. In Fig. 10b, however, which shows a very small flake in the same material, the inclusions that permitted the separation of the metal to occur and determined the course that the flake took are very evident.

Microstructure

When a section of flaky steel at right angles to the general plane of the flake is examined after properly polishing and etching, it is seen that the metal bordering the flake is normal in its structure up to the immediate face of the separation; that is, the flake has no depth. Figs. 11 and 12 show the structure of a chrome-nickel steel and a plain nickel steel, as

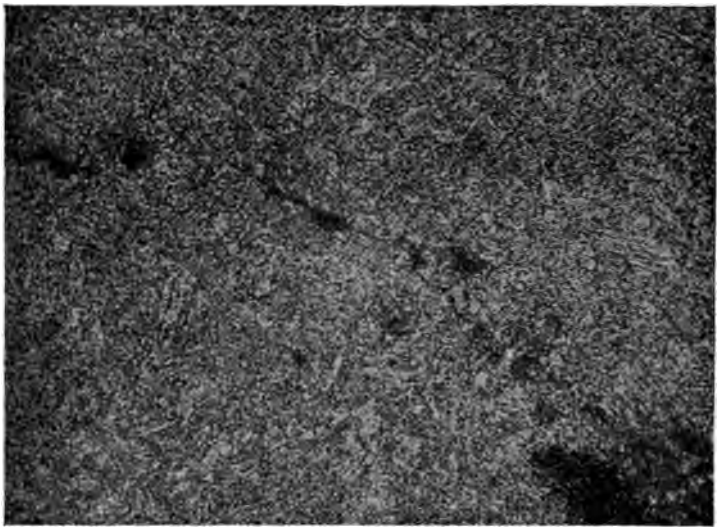


Fig. 11.—Chrome-nickel steel forging showing section through flake; structure is normal up to face of flake. Two per cent. alcoholic nitric acid. × 500.

related to a flake extending through the piece. The flake appears as a discontinuity in otherwise normal material. Fig. 13 shows a section of the same specimen after quenching; the metal is hardened uniformly throughout, the coarsely crystalline appearance noted on the face of the flakes in such material, Fig. 3, is a surface configuration only—the real grain of the metal constituting the flake is refined to the same degree as elsewhere in the specimen. The discontinuity in the metal, which in the fracture appears as the flake, is an intercrystalline one. Fig. 14, which shows the structure of a billet of chrome-nickel steel before receiving any heat treatment whatever, shows how closely the course of the

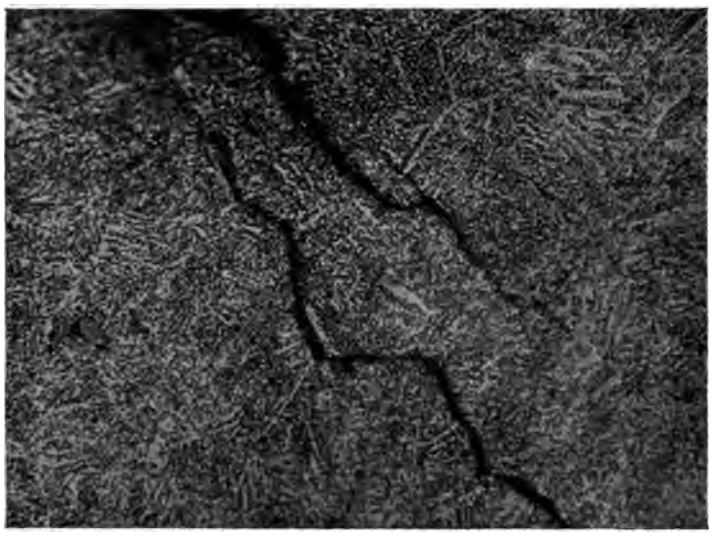


FIG. 12.—NICKEL-STEEL GUN FORGING, HEAT TREATED, SHOWING SECTION THROUGH A FLAKE, C, 0.37; NI, 2.83; CR, 0.01. MICROSTRUCTURE OF METAL IS NORMAL UP TO IMMEDIATE FACE OF FLAKE. ETCHING, 2 PER CENT. ALCOHOLIC NITRIC ACID. × 500.

FIG. 13.—MATERIAL SAME AS FIG. 12 AFTER HARDENING. ETCHING, 2 PER CENT ALCOHOLIC NITRIC ACID. × 500.

flake follows the grain boundaries, which here are plainly shown by the different orientation of ferrite planes within the different crystals. The

FIG. 14.—SECTION OF CHROME-NICKEL BILLET AS ROLLED SHOWING INTERCRYSTALLINE FLAKES. ETCHING, 2 PER CENT. ALCOHOLIC NITRIC ACID. × 500.

FIG. 15.—NICKEL-STEEL GUN FORGING, SAME MATERIAL AS FIG. 4, SHOWING APPEARANCE OF FLAKES IN MATERIAL AS POLISHED BUT UNETCHED. × 100.

presence of flakes in this billet was first demonstrated by polishing a cross-section of the billet. A series of interior radial cracks was found which, when broken open, had the characteristic appearance of flakes.

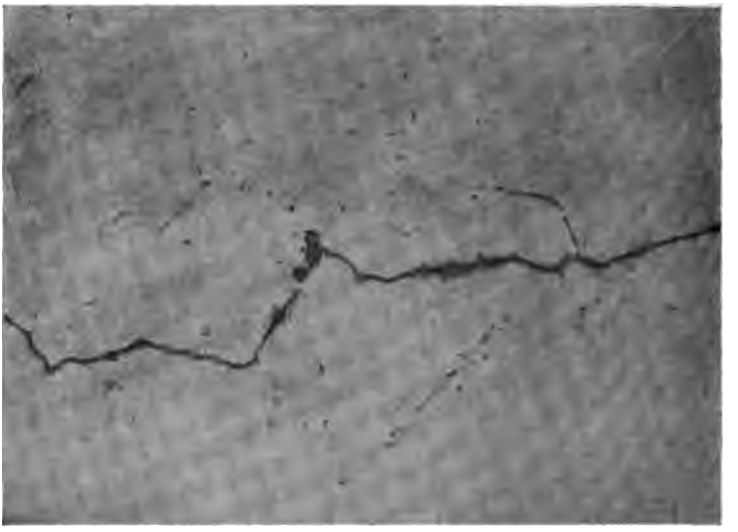
The intercrystalline nature of the separation naturally leads to the examination of the material for the presence of intercrystalline inclusions or other features, which would account for the separation taking place in this manner. The examination must be restricted to very small flakes or to the extreme edges of those of larger size. Fig. 15 shows the appearance of a flake at a moderate magnification—the flake appears merely as a crevice or discontinuity in the metal. It is best to limit the examination at higher magnification to the fine hair-like cracks or extensions of flakes at the ends of the wider openings, as is indicated by the arrows. Fig. 16 shows such a crack, which in reality consists of an extremely thin film of slag (magnification 500 diameters). Figs. 17a and b show sections of the film in two other similar flakes. The film is often discontinuous and shows breaks in it; in Fig. 18, it occurs as a continuation of isolated globules in chrome-nickel steel, carbon 0.39, nickel 3.08, chromium 0.87. Fig. 19 shows a flake the course of which is outlined by slag, the globules of which are much larger. In general an examination of the more open part of the flake shows little that is suggestive. Occasionally, however, isolated globules of glassy slag may be detected within the opening, as in Fig. 20. By deep grinding so that the outer margin of the flake is sectioned, particularly if the section is slightly oblique to the plane of the flake, traces of a glassy slag may be found in the cavities, as in Fig. 21. The wide portions of the crevice, on the whole, however, are not found to be completely filled with a continuous slag film.

Upon etching the specimens showing slag films with hot alkaline sodium picrate, it is noted that, in general, the gray color of the film is not changed. The work of Comstock[1]

[1] G. F. Comstock: *Trans.* (1916) **56**, 553–560.

FIG. 16.—NICKEL-STEEL FORGING, EXTREME END OF FLAKE INDICATED BY ARROWS IN FIG. 15 IS SHOWN TO CONSIST OF AN EXTREMELY THIN FILM OF SLAG. UNETCHED. × 500.

a



b

FIG. 17, *a* and *b*.—NICKEL-STEEL FORGING, SAME MATERIAL AS FIG. 12, SHOWING SLAG FILMS ENCLOSED WITHIN THE FLAKE. UNETCHED. × 500.

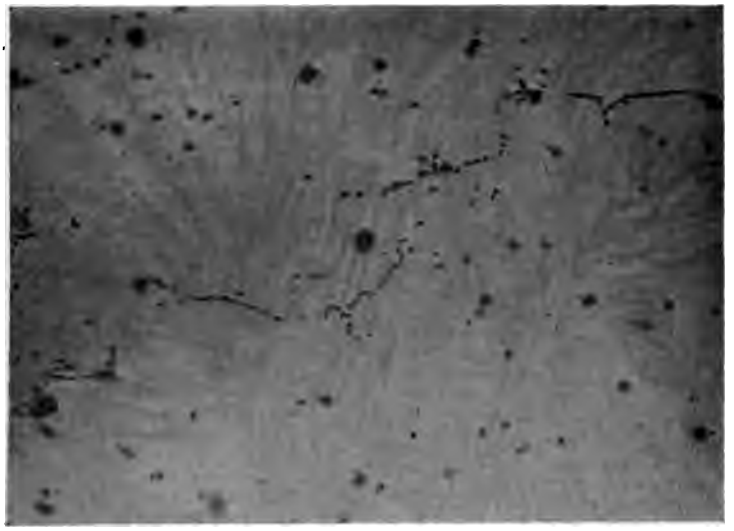


FIG. 18.—CHROME-NICKEL STEEL FORGING, MATERIAL SIMILAR TO FIG. 11, SHOWING TINY SLAG INCLUSIONS WITHIN THE FLAKE. UNETCHED. × 100.

FIG. 19.—CHROME-NICKEL STEEL FORGING, SAME AS MATERIAL IN FIG. 19, SHOWING FLAKE OUTLINED BY SLAG INCLUSIONS. UNETCHED. × 500.

FIG. 20.—NICKEL-STEEL FORGING, SAME MATERIAL AS FIG. 4, SHOWING SLAG GLOBULES ENCLOSED WITHIN WIDER PART OF CREVICE OF FLAKE. UNETCHED. × 500.

FIG. 21.—NICKEL-STEEL CONTAINING FLAKES, SAME MATERIAL AS FIG. 12, ING TRACES OF SLAG ENCLOSED WITHIN CREVICES OF FLAKE. ETCHED WIT ALKALINE SODIUM PICRATE. × 500.

has shown that this is an indication that such inclusions are of the nature of true slag rather than sulfide.

Probable Nature of Flakes

The results of every kind of examination of flaky steel indicate that flakes are discontinuities within the steel. These discontinuities are often found to be associated with extremely thin films of slag which are either continuous or formed of tiny isolated globules. The appearance of the face of the flake to the unaided eye or when examined with a hand magnifier suggests strongly that the defect had its origin early in the history of the metal. The coarse crystals have rounded corners and faces and in general the appearance suggests plastic bodies that have been

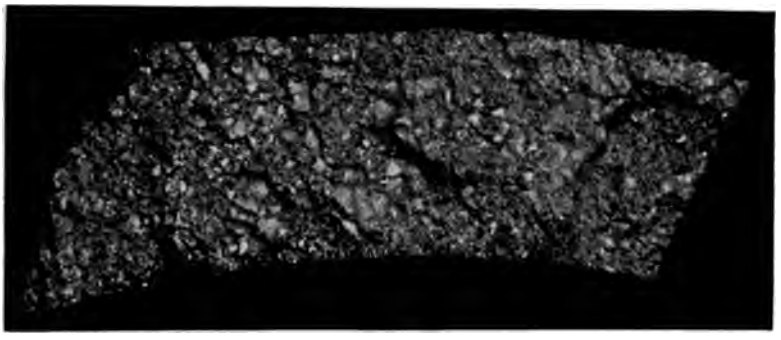


Fig. 22.—Fracture of defective chrome-nickel steel showing conditions within ingot. Sample was from 30-in. ingot about 5 in. from center; natural size.

squeezed together and failed to adhere. There is no trace of the glistening cleavage faces of crystals, which are always to be found when a coarsely grained metal breaks normally, that is by an intracrystalline break. The usual symmetrical arrangement of the flakes, except in forgings of which parts have been very much displaced, such as the twisted portions of crankshafts, also suggests that they were formed early in the history of the piece. The fact that the flakes are all intercrystalline in their nature also confirms this. The sectioning of large-sized blooms and of ingots has amply verified the conclusions based on observations of the finished forgings and demonstrated that the defect originates in the ingot and persists throughout the entire forging period and the subsequent heat treatment.

Although flakes have been found to be associated with slag inclusions, either as thin films or as isolated globular masses, the fact that steel is "dirty" is by no means a sure or safe criterion for condemning it as "flaky." Without doubt, however, the presence of slag inclusions, which later may be squeezed down to the form of films, is a condition

that aids very materially in permitting a separation of the metal to occur and also assists in the prevention of the subsequent welding up of such interior cavities. While inclusions and slag films have been found associated with flakes in a large number of specimens of flaky steel examined, in a considerable number of samples such conditions were not found. Particularly was this true of many of the nickel-chromium steels examined. Hence no general statement can be made that the presence of

a

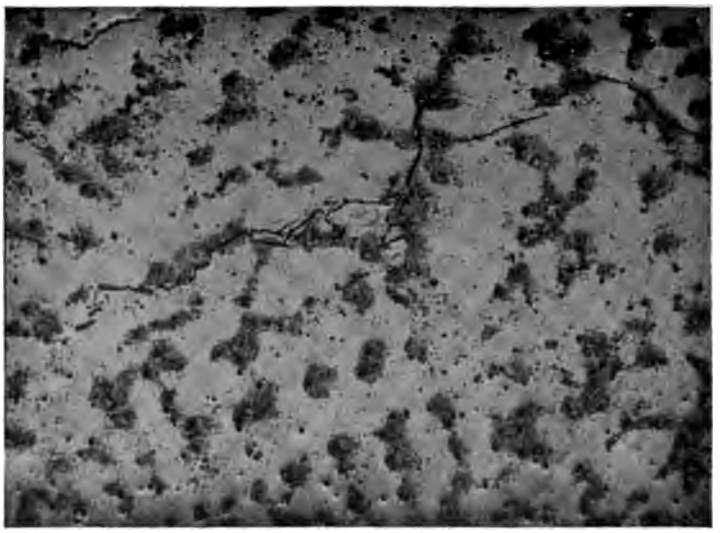


FIG. 23.—*a*, SECTION OF MATERIAL OF FIG. 22, ETCHED WITH 2 PER CENT. COPPER-AMMONIUM-CHLORIDE SOLUTION TO SHOW THE DENDRITIC STRUCTURE AND SHRINKAGE CRACKS; NATURAL SIZE. *b*. SAME MATERIAL, × 15. INTERIOR CRACKS, WHICH GIVE RISE TO FRACTURE SHOWN IN FIG. 22, OCCUR IN BETWEEN THE DENDRITES WHERE THE SLAG INCLUSIONS ARE LOCALIZED.

such inclusions is a necessary condition for the formation of flakes. This condition, however, if it existed would aid very materially in their formation.

In Fig. 22 is shown the appearance of a specimen taken from the interior of an ingot of defective chrome-nickel steel of the composition, carbon 0.40, chromium 0.35, nickel 2.75. The condition revealed in the fracture without doubt represents the initial or ingot condition of

the defect, which later in the history of this piece would be designated as flakes. The dull smooth coarsely crystalline spots in the fractures have the appearance of metal that has failed to unite to the adjacent metal. A section through the specimen is shown in Fig. 23; the coarse dendritic structure characteristic of the cast metal is clearly shown, also some interior crevices continuous with the smooth surface areas of the fracture. These cracks occur between the fingers of the dendrites, which is the portion of the metal that is the last to solidify and contains most of the inclusions (Fig. 23b) and hence is the weakest part of the steel. The cracks have the same appearance as the shrinkage cracks often found in defective castings of bronze and brass. Such shrinkage cracks invari-

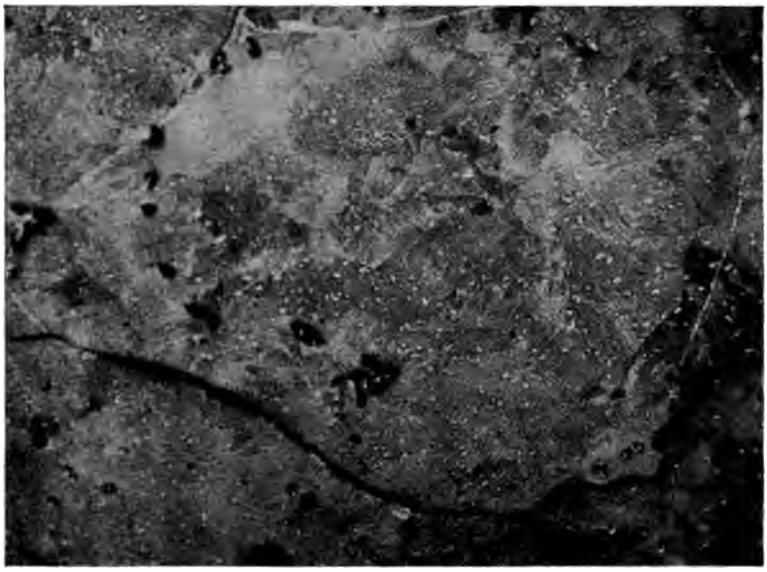


FIG. 24.—SAME MATERIAL AS FIG. 22; INGOT-CRACKS ARE LARGELY INTERCRYSTALLINE. ETCHING, 2 PER CENT. ALCOHOLIC NITRIC ACID. × 50.

ably form in the crevices between the branches of the dendrites and so far as can be judged from their microscopic appearance these ingot cracks in steel should be considered as of similar origin. Fig. 25 shows how these shrinkage cracks often pass through slag globules of considerable size; such globules, upon forging the metal into shape, may result in the thin films shown in Figs. 16 and 17. Fig. 24 illustrates the intercrystalline nature of these ingot cracks.

Another type of defect in steel, very closely related to flakes, is the so-called "silver streaks;" Fig. 26 illustrates the appearance of these in a chrome-nickel steel. Each white elongated patch shows in its center a tiny thread of greenish slag and the surrounding metal has the appearance, as in a true flake, of never having been in direct union with the op-

posing metal. Other spots, larger than the silver streaks shown, containing rather prominent slag plates have often been noted in the faces of fractures of steel. The metal underlying and surrounding such slag plates has the bright clean silvery appearance found in the flakes. Such

FIG. 25.—SAME MATERIAL AS FIG. 23. SLAG GLOBULE LYING IN COURSE OF CRACK, UPON FORGING STEEL, WOULD PROBABLY RESULT IN FILM-LIKE ENCLOSURES SIMILAR TO THOSE OF FIGS. 16 AND 17. UNETCHED. × 100.

spots are usually designated as crystalline breaks due to the presence of slag. It appears very probable that such defects are due entirely to rolled out slag plates and that the presence of initial shrinkage crevices or ingot cracks is not necessary for their formation. Defects of this type are usually revealed in steel that is tested transversely; that is, the specimen is taken so that its axis is perpendicular to the fibers, due to forging, instead of parallel as is the usual case. Flakes, however, are found in both transverse and longitudinal sections; in chrome-nickel steel, in particular, they are found to extend in all directions in the ingot. In the plain nickel steels they are usually more symmetrically arranged.

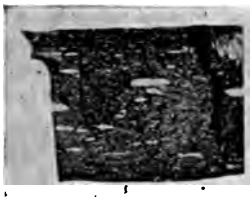


FIG. 26.—FRACTURE OF CHROME-NICKEL STEEL FORGING SHOWING DEFECT OF SILVER STREAKS; EACH WHITE ELLIPTIC AREA CONTAINS A SHORT THREAD OF GLASSY SLAG. NATURAL SIZE.

On the whole, the microscopic appearance of flaky steel leads to the following conclusions regarding the nature and origin of flakes. They originate in the ingot, in which state they have the appearance of intercrystalline shrinkage cracks. They occur in the cast metal along with the slag inclusions in the angles between the branches of the tree-like den-

dritic crystals. They persist throughout the history of the forging into the finished form as discontinuities in the metal, often associated with slag films, which have resulted from the working down of the slag inclusions with which they are associated from the beginning. The coarsely crystalline appearance of the flake is a surface configuration only and is a record of the crystalline condition, that is, the dendrites, existing in the steel at the time the discontinuity originated. Though they are usually associated with dirty steel, the converse is not true; the fact that steel is dirty or badly contaminated with slag inclusions is not a sure criterion for condemning the metal as defective because of flakes.

DISCUSSION

(Includes also the paper by Messrs. Clayton, Foley and Laney, pp. 211–245.)

P. E. McKINNEY,* Washington, D. C.—Observation made after examination of a number of defective gun forgings, together with a history of their manufacture, has strongly suggested that so-called "flakes" and woody or fibrous fractures are not necessarily analogous and are often caused by entirely different conditions. While some flaked bars may show fibrous areas in the fracture, it does not necessarily follow that the conditions causing these areas are the same as cause the fibrous, or so-called laminated, structures so commonly met in gun forgings subject to transverse tensile test; the fibrous structure formed in some flaky bars are merely incidental to the conditions causing the flakes.

It is believed that the so-called flakes, or bright silvery spots, are mainly due to inherent physical defects in the steel, which existed at the very origin of the steel, rather than metallographic defects caused by subsequent irregularities in mechanical or thermal treatment. While great care in forging and treatment may, in a measure, tend to overcome the evil effect of the defects, with many heats of ingots, no such precautionary measures are needed to produce perfect forgings and with such heats of steel it would be impossible, no matter how carelessly handled, to produce intentionally a forging containing the defects commonly known as flakes.

It has been found possible, on several occasions, to actually predict flaky forgings from an inspection of the ingot after it had solidified, and in such cases no amount of careful forging or heat treatment would prevent the occurrence of such defects in the forging. The inspection of the physical condition of the ingot, and especially the character of the shrinkage on the top of the ingot, has been given too little thought and study in this connection.

Examination of Figs. 46, 47, and 48 of the paper presented by Messrs. Clayton, Foley, and Laney would lead most strongly to one conclusion;

* U. S. Naval Gun Factory.

namely, that the steel contained gas cavities or blowholes, and it is safe to say that had this specimen been similarly treated while still in the ingot form, the defects would have been discovered at that time. Defects of this character are not new in forging steel but are inherent to all so-called "open steel." Their presence would not be appreciably noticed in longitudinal test specimens or in large hollow forgings that can be worked to an appreciable extent on a mandrel, as, with longitudinal work, the defects are drawn out into threads, which are not noticeable unless fracture occurs at right angles or other than directly parallel to the major axis of the forging. While it might be stated that serious flakes have been encountered in longitudinal tests on crank-shaft forgings, a careful study of the forging practice followed in the production of such forgings will show that almost invariably the working of the forging in the die or tools has tended to slightly upset the grain. If commercial practice required the transverse test on carbon-steel forgings to as great an extent as ordnance work requires such test on nickel-steel forgings, it is believed that flaky steel would be found almost equally prevalent in carbon steel.

Mr. Rawdon's statements that flakes have their origin in the ingot can be substantiated by the findings of other investigations, as in numerous cases of suspected and known flaky steel the defects could be discovered in the ingot before any forging or heat-treatment work had intervened to complicate conditions.

The statement has frequently been made that flakes are characteristic of steel made by certain processes, but this can hardly be substantiated by results of practice under different processes throughout the country, as it is a well-known fact that many mills manufacturing steel by acid open-hearth, basic open-hearth, basic electric, or other processes rarely encounter any trouble while others manufacturing under the same practice encounter continuous trouble from this score. Since it has been conclusively shown that the trouble cannot be attributed to poor quality in so far as composition is concerned, the logical conclusion to be drawn is that either faulty melting and pouring practice taken individually or collectively must be the cause of the trouble.

A study of the complete history of melting heats in a great number of cases of satisfactory and unsatisfactory steel has shown that steel that is well deoxidized and degasified before tapping from the furnace and is poured at the proper temperature will not produce flaky conditions, and that ingots not produced under these conditions might economically be consigned to the scrap yard before putting a lot of forging labor on them. It is not the purpose of the discussion to go into the details of methods of operation necessary to produce perfect steel, but rather to call attention to the fact that so-called flakes are not inherent to all steel produced under certain conditions, and that if the practice followed by steel

producers who do not encounter trouble was closely followed the present difficulties would be materially reduced.

S. P. ROCKWELL* and B. E. FIELD,† New York, N. Y. (written discussion‡).—The papers and the previous discussion on the so-called flakes have connected them chiefly with ingot and forging structures. Our experience has shown that flakes also occur in the rings or cups in ball-bearing work. The existence of these has been known for the past six years. The rings in which the flakes occur are made from open-hearth steel of the following analysis: Carbon, 0.10 to 0.20 per cent.; manganese 0.25 to 0.50 per cent.; phosphorus, not over 0.04 per cent.; sulfur, not over 0.04 per cent.; silicon, not over 0.20 per cent.; chromium, 0.65 to 0.85 per cent.; vanadium, 0.02 to 0.14 per cent. They are made by machining from cold-rolled bars or by cupping flat stock and cutting out the bottom of the cup. The rings are cemented from all sides at 1675° to 1700° F. (912° to 922° C.) so that the carbon content of the core is increased to 0.50 to 0.70 per cent. while the outside is eutectoid in composition. Proper heat treatment after air-cooling from the cementing temperature refines the whole so that it has all the appearances of a high-grade tool steel. The first quench is from 1560° F. (849° C.) in oil and the second from 1430° F. (777° C.) in oil.

The fracture of such a ring, when a so-called flake is present, appears as a metallic white granular area similar to the so-called flakes in gun forgings. These flakes are accompanied, in a majority of cases, by a straight line across the center of the flake parallel to the axes of the ring. This flake has never been found to intercept the surface of the ring, whereas the line referred to does extend to the surface.

These flakes have no thickness and therefore can not be polished and examined under the microscope. Examination of the line under a magnification of about 10 diameters shows a greenish-gray discoloration somewhat suggesting ingot gases. Breaking the ring again near the flake will give a perfectly refined grain structure, providing, of course, another flake is not present at that particular point. On grinding off the face of the ring and polishing, radial cracks can be seen. These cracks never extend to the edge of the polished surface; they are not symmetrically distributed around the ring and the number has been found to vary from none to 29. Breaking the ring through such a crack always shows a flake. Breaking between cracks shows a good structure. No difference has been found in the occurrence of the flake whether the break was caused by tension or compression.

Inasmuch as the cracks—for such the flakes seem to be—have occurred in rings made from bar stock and from rings made from flat stock, and also as the cracks never intersect the surface of the ring, it is difficult to see how they can be in any way related to the ingot or forging structure.

* Captain, Ordnance, U. S. A. † New Departure Mfg. Co. ‡ Received Feb. 18, 1919.

That the flakes or cracks are present in the rings directly after the cementing operation and previous to the hardening treatment is shown by the lack of strength exhibited by some of the rings and the phenomenon of the cracks appearing on rings polished off perpendicular to the axis. The flake of the fraction of the cemented ring cannot be seen clearly as such at this point because of the coarse grain structure of the entire piece. No flakes have ever been detected in the rings previous to the cementing of the work. We must conclude, therefore, that the hardening treatment is not to blame and that these rings cannot be considered as true fire cracks.

Another interesting phenomenon that occurs in high-carbon chrome tubing used in ball-bearing work is that of two parallel surfaces extending toward the center of the tube at an angle and showing on the outside as two parallel lines running spirally along the length of the tube. This phenomenon of the angle and spiral formation is due apparently to the action of the piercing rolls when the steel is in a plastic condition. The steel between these two surfaces is low in carbon and shows white after hardening in oil, whereas the adjoining steel is black. Rings made from such tubing have no strength and a fracture at either of these surfaces has a usual flaky appearance. In this case ingot structure might account for the trouble.

Further, motor tappet rods that were hot-headed showed, after cementing, the so-called flaky structure running spherically about the head. In this case, thin sawed sections allowed the so-called flake to be punched out, showing that it had volume. A tubular seam high in impurities was detected in the length of this steel by sulfur prints and etching with Stead's reagent and the heading operation undoubtedly gave the volume to the flake by forge shortening.

Three phenomena have been cited giving rise to the same flaky appearance in the fracture and yet no two of them seem in any way connected. Are we all discussing the same kind of a flake or are there several kinds manifested in different ways?

J. A. MATHEWS,* Syracuse, N. Y. (written discussion†).—The two very interesting papers submitted constitute about the entire literature of this subject. The matter is one that has received a great deal of attention during the past year or two, although the defect itself is not entirely new in steel manufacture. Nevertheless, it does not seem to have received much attention so far as published articles are concerned. The experience of the authors of the papers seems to be most largely in connection with heavy forgings, and for the most part with plain nickel steel, whereas the material that I have had an opportunity to examine has been mostly of the higher types of chrome-nickel steel used in the form of 4 by 4-in. (10 by 10-cm.) billets, or in forgings made therefrom.

* President, Halcomb Steel Co.

† Received Feb. 18, 1919.

About a year ago, very serious trouble was developed in connection with the aircraft production work, and in June of last year a conference of nearly 100 representatives of our own and foreign governments, together with makers of steel, forgings, and aircraft engines, discussed this matter, and it was at that conference that I saw the first examples of this defect. I have never seen it in our own product. As an outcome of the conference, a Committee on Aircraft Engine Forgings was formed, composed of several steel makers and representatives of the Aircraft Production Department. The writer had the honor of serving as Chairman of this Committee. Unfortunately for the Committee, but fortunately for the aircraft industry, the defect seemed to automatically disappear about as soon as attention was called to it formally. A questionnaire submitted

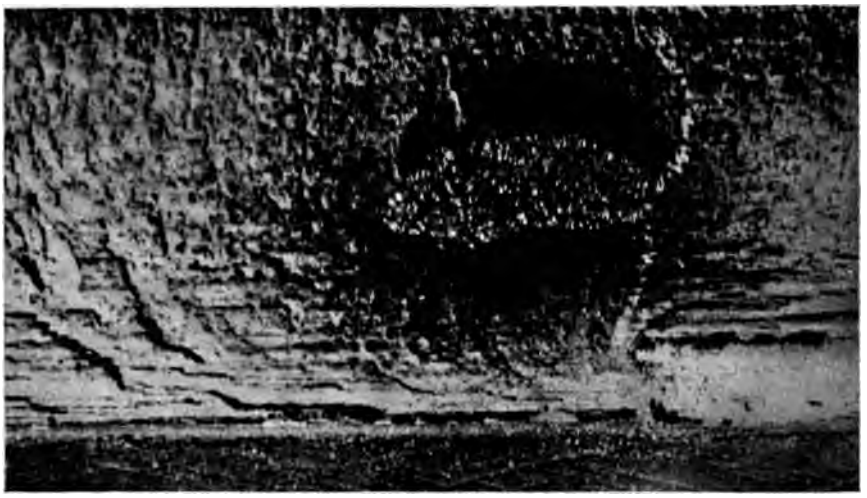


FIG. 1.—SNOWFLAKE IN 4-IN. ROUND BAR, CHROME-NICKEL STEEL (C, 0.32 TO 0.40 PER CENT.; CR, 0.50 TO 0.80 PER CENT.; NI, 3.25 TO 3.75 PER CENT.).

to a number of representative makers of alloy steels, forgings, and engines, showed that many of them were not at all familiar with the defect known as "snow flake," and those who had seen the defect reported it as occurring in chrome-nickel steels, although a few said they had seen it in nickel steel, and possibly in some other types. There has been some question in my mind whether the defect described by the authors of these papers, and as experienced by the gun makers, was actually the same as that which we studied in connection with smaller billets and forgings.

I note in both papers that the condition is considered as arising in the ingot, and in some cases internal cracks or fissures have been noted in the ingot itself, which might persist throughout subsequent operations. I am not convinced, however, that all of the defects found in the finished article started from cavities in the original ingot. It appears to me that

cavities or cracks that might exist in the ingot, which did not weld or close in subsequent operations, would present, on fracture, a rubbed appearance striated in the direction of working; in fact, would present more of the appearance of the silver streaks referred to by Mr. Rawdon. It does not seem quite reasonable to suppose that such crystalline surface could persist during the operation of forging or rolling without being pretty well obliterated and streaked in the direction of the flow of metal. Furthermore, we were furnished a defective propeller shaft forging, which was 4¼ in. (10.7 cm.) in diameter, and on one end had a heavy upset flange 10 in. (25 cm.) in diameter. This material showed flakes from end to end, but they were less apparent in the heavy upset portion than in the round shaft. A part of the round shaft was cut off and reworked to 2 in. round,

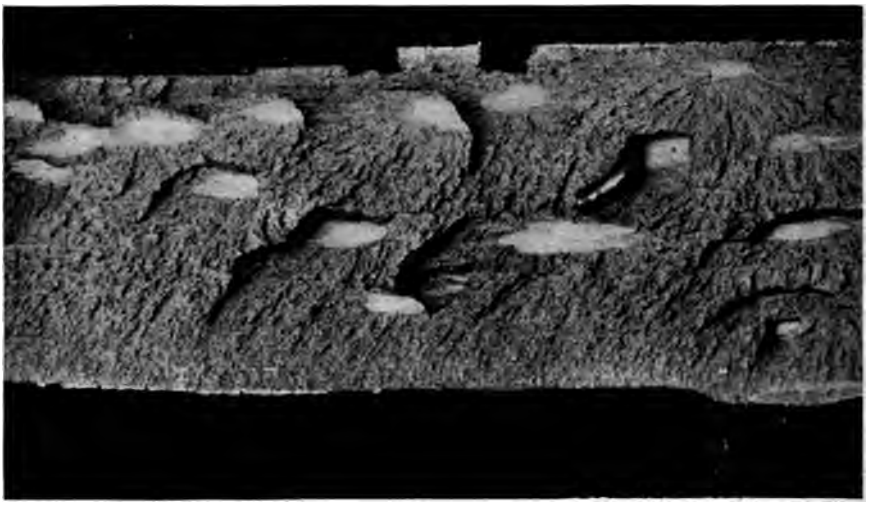


FIG. 2.—SILVER STREAKS IN 4-IN. ROUND BAR OF CHROME-NCKEL STEEL (C, 0.32 TO 0.42 PER CENT.; CR, 0.50 TO 0.80 PER CENT.; NI, 3.25 TO 3.75 PER CENT.).

and in this condition all signs of flake disappeared; neither could they be found in the natural fracture of the steel, nor after hardening, when examined in either a lengthwise or transverse fracture, and a microscopic examination failed to show any traces of them. It should be borne in mind that in dealing with the higher types of chrome-nickel steel, we are dealing with a steel that is semi-airhardening and very sensitive to heating operations.

In connection with the aircraft work, we found that only two or three mills seemed to furnish such material, and that these mills had not had great experience with this type of steel. It is a steel that requires great care in the handling at every stage from the melting down. Excessive casting temperatures, and excessive forging and rolling temperatures, predispose the metal to the formation of flakes, even though they may not

exist as actual fissures in the original ingot. It is what would be called in medicine an "hereditary tendency."

The experience of the authors, in reference to the normal appearance of the metal right up to the edge of the flake, conforms with our experience, yet we cannot quite agree with Mr. Rawdon's statement that, in cases where some slag is found included in the fissure, the metal is, or necessarily should be, normal. It is our experience that metal in contact with the slag inclusion generally shows some structural change, usually due to decarburization. Mr. Rawdon states that the snow flakes are usually symmetrically placed with reference to the axis of the bar or forging. This we have found to be generally true, but occasionally we have found flakes at right angles to one another. The appearance of the flake to the eye, or with a low-power glass, is so crystalline that we have felt that these might occur in steel after all hot work on the material had been completed; that is, they might constitute a series of internal bursts, possibly filling what were the original spaces between the primary dendrites in the ingot.

It has impressed me as very strange, in hardening portions of billets containing flake, that the flakes did not open up in the hardening but seemed to remain the size that existed in the unhardened piece; that is, one would expect that the contour of the flake in the natural condition would extend in hardening so that two different crystalline appearances would be noted, one indicating the appearance of the flake as it existed before hardening and the other due to the hardening crack resulting therefrom in all directions and surrounding the original flake. So far, I have not observed any tendency of these flakes to extend in dimensions in the hardening operation.

I am submitting herewith two photographs, one showing characteristic snow flake and the other so-called silver streaks. These photographs are only slightly enlarged over the original. In all that I have said, I have been referring to chrome-nickel steels, and these specimens are chrome-nickel steel of approximately 0.40 per cent. carbon, 3.00 per cent. nickel, 0.80 per cent. chromium, and 0.50 per cent. manganese.

O. A. KNIGHT,* State College, Pa. (written discussion†).—I have read with great interest the papers by Mr. Rawdon and by Messrs. Clayton, Foley and Laney, having also conducted a number of investigations regarding the cause of flakes in steel.

The characteristic flakes in steel are beautifully illustrated by the photographs Mr. Rawdon has presented. Fig. 1 is an excellent example. A very interesting feature of this work is that a much larger flake is produced through fracturing the steel by alternating stresses than by

* Instructor, Metallurgy and Metallography, Pennsylvania State College.
† Received Feb. 18, 1919.

impact. This can be explained as follows: the flaw is a discontinuity in the metal to begin with and by repeated stresses the discontinuity or cavity is gradually enlarged along the planes of weakness until final rupture occurs suddenly through the remaining metal. This results in the growth or enlargement of these flaws, a phenomenon that is not permitted to take place when the fracture is a result of sudden impact.

The radiographs are both interesting and instructive, but would not other defects and discontinuities in the metal also be shown by this method of examination which would render it more or less difficult to distinguish a flaw from other defects?

In Fig. 6 of Mr. Rawdon's paper, there is shown a peculiarity to which I wish to call special attention. Macroscopic etching has here shown these flaws to extend in a radial direction. I have investigated a number of gun forgings of a similar nature and, in all cases examined, the

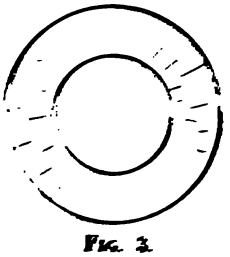


FIG. 3.

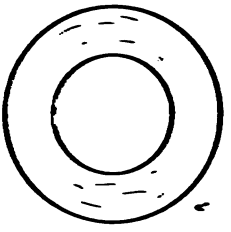


FIG. 4.

DIAGRAMMATIC ARRANGEMENT OF FLAKES.

direction of the flakes was radial, as shown by Mr. Rawdon. This being true, how can we conclude that these flakes are a result of forging? Or, again, how can we connect this fact with the hypothesis set forth in the paper by Messrs. Clayton, Foley, and Laney, in which they attribute the cause of flakes to a fusion of the constituents that have the lowest melting point, after which these parts are forged out flat? To further illustrate this point I have drawn two sketches. Fig. 3 shows the flakes extending in the radial direction, in the same manner that they actually occur. If these flakes were caused by forging, or if they were spots of fused metal forged flat, they could not possibly extend radially but would be tangential or circumferential as illustrated in Fig. 4. Therefore it is impossible for them to be the result of forging out such fused particles as has been suggested, and at the same time extend radially.

A point little touched upon in either paper is the possibility of producing flakes during the heat-treating process. This I have investigated, in cooperation with F. F. Hansen, general metallurgist for the Bethlehem Steel Co., Bethlehem, Pa., and typical flakes were successfully produced. This experiment consisted in heating a gun tube (having a composition similar to those under discussion) very rapidly to the quench-

ing temperature, quenching until cold in water, and then heating quickly to the drawing temperature of 650° C. followed by rapid cooling to atmospheric temperature. The tube, which before this treatment did not show this defect, was then examined for flakes and numerous large flakes were found on longitudinal radial fractures. Some of these were 0.75 in. (19 mm.) in length by from 0.25 to 0.5 in. in width. This points to the formation of flakes being due to internal stresses induced in the metal by rapid heating and cooling. Polishing a transverse section and etching for microscopic examination showed the flakes to run in a radial direction. This is best shown by etching with concentrated hydrochloric acid. The action of the acid is greatly stimulated by heating but very good results can be obtained by the application of the acid to the cold surface of large pieces if sufficient time is allowed for the action to take place. Boiling for 15 min. usually gives excellent results. Some investigators prefer to use a solution of sulfuric and chromic acid rather than hydrochloric acid, on account of the fumes of the latter.

I have also examined a great number of flakes microscopically and found, as did Mr. Rawdon, that a flake has the appearance of a discontinuity in the metal. However, these flakes occur in steel that contains both coarse and fine grains, such for example as are shown in Figs. 8 and 10 in the paper by Messrs. Clayton, Foley, and Laney. Also, the structure here shown of sorbite with ferrite in the cleavage planes is commonly found in flaky steel. I have not observed such evidences of overheating to accompany the occurrence of flakes as have been shown in the paper. I was successful in removing these flakes by forging. A section of a gun tube that contained many flakes, all of which extended in the radial direction, was cut in two and forged. At first only a slight reduction of area was made, after which the steel was polished, etched, and examined microscopically for evidences of flakes. The original flakes were still present but their direction had been changed and followed the lines of flow of the metal in forging. A second forging resulted in the complete elimination of them. This would be expected if flakes are discontinuities in the steel with bright, unoxidized surfaces. The flakes could not be found either in the fracture or by etching the polished surface.

It is my opinion that flakes, as we know them, are the result of stresses set up within the metal during heat treatment and their location is determined by planes of weakness already existing in the metal prior to this heat treatment. Sufficient data have not yet been collected to prove this, but the evidence points to this conclusion.

H. D. HIBBARD, Plainfield, N. J.—I would like to ask Mr. Foley whether the steel was open-hearth or electric?

F. B. FOLEY.—This steel was mostly electric.

H. D. HIBBARD.—I would like to know something about the steel before it was cast and something about the ingots and their treatment.

F. N. SPELLER, Pittsburgh, Pa.—It might help if we were to emphasize the effect of the welding quality of the steel on the heating of the cracks. If these flakes are due to cracks and the cracks are due to the condition of the forgings and the strain in some previous process, we would expect these cracks just as much in carbon steel as in alloy steel and would probably get them if it were not for the superior welding quality of that steel. I have made a particular study of the factors that have to do with the welding quality, and those alloys, as we all know, have a very marked effect, but it may not be common knowledge that less than 10 per cent. of nickel very materially reduces the welding quality of the steel. It is a very different matter, however, with these alloyed steels, especially nickel steels.

As to the position of those flakes in the ingot, may not that have something to do with the position of the ingot? They may occur in the lower third of the ingot or in the upper half, or inthe center. That would constitute a strain wherever those weak points happen to be.

J. E. JOHNSON, JR., New York, N. Y.—Are these defects a matter of recent development and is that because they are gun steels only; are they common in some particular composition of steel and relatively uncommon in others? I have not heard that brought out and I think it would be very interesting if we could know more about the history of this subject and the kind of steels in which they occur. Is the reason that this thing comes to the front now simply that we have been trying to make guns and have run into this difficulty in trying to make gun steel?

A. A. STEVENSON, Philadelphia, Pa.—I see some men around here who I think are just as much interested as I am in one phase of this question. Mr. Hibbard touched on it a little. We have heard a good deal about what flakes are and we have heard something about where they are, but some of the people who make steel would like to know why they are. The flakes are, I think I can say, most noticeable in nickel steel and in basic steel, and not so noticeable in acid steel. I think for 6 or 9 months after we started in the manufacture of gun steel, we didn't have a flake. We had a number of gun manufacturers at the plant last summer and they spent their time looking over test pieces to find flakes, but they couldn't find any. Whether the disease was catching or not, I don't know, but it wasn't long after they were there that we began to have flakes, and any quantity of them. We had made a radical change in our practice. When we started to make gun steel we used low-phosphorus scrap, but we accumulated a lot of nickel-steel

scrap in turnings and later our charge was made up almost exclusively of low-phosphorus pig iron, nickel scrap, and nickel turnings.

In answer to Mr. Johnson's question as to the interest in this subject at the present time, I would say that owing to the war, the Government has required large quantities of nickel steel and nickel-chrome steel for ordnance purposes. The number of rejections due to flaky steel caused representatives of the Ordnance Department and the manufacturers to devote a great deal of time and study to the subject. Experience has shown that nickel and nickel-chrome steel are especially prone to flakes.

J. S. UNGER,* Pittsburgh, Pa.—The question of snowflakes, as stated by Doctor Mathews, was brought up about the middle of last year. It is true that quite a number of the members of that committee believed that the term "snowflake" was new; snowflake could not be considered strictly a metallurgical name. At that meeting it was called by some "lemon spots" and we have known it, in some of our work, as "goose eggs." Now if snowflake, lemon spot, and goose egg are the names for the same disease, I want to say that I have known of snowflakes for at least 20 years.

My experience would indicate that snowflakes are most common to steels that contain chromium, although I have found them in nickel steel, and quite recently have found them in steel containing about 0.25 per cent. of silicon without any nickel or chromium. In making a study of the snowflake situation, I don't think one can attribute it to the shape of the ingot. I have found it in square ingots, rectangular ingots, round ingots, octagon ingots, and fluted octagon ingots. Nor can it be attributed to the method of casting. I have found it in ingots that were top cast and bottom cast, in iron molds and in sand molds, with a hot top and without a hot top. The size doesn't influence it, for I have found it in ingots 10 by 10 in. and 40 by 120 in.

As regards position, in a general way I believe snowflakes are more prevalent near the top of the ingot than at the bottom, although I have found them in both places. I have found them at what represents the center of the ingot at the middle portion, and at the edge. I have found snowflakes in ingots that have been heated all over at once, by being charged in the soaking-pit, and in ingots in which the tong hold extended outside of the furnace. I have also found snowflakes in ingots in which there was an excessive amount of reduction, let us say a 25 by 25-in. ingot reduced to a 4 by 4-in. as compared to a 25 by 25-in. ingot reduced to a 14 by 14-in. The peculiar feature about it is that the same sized ingots, cast one right after the other, but sent to forgeman A and forgeman B, will show a different number of snowflakes. I have found that

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one man would not report any and the other would report an abundance of snowflakes.

Recently I had occasion to examine some steel which was of about 0.25 per cent. carbon and about 0.3 per cent. silicon. On making a tensile test on a piece 1¼-in. octagon, I found a snowflake right in the center. I took the same bar and bent it until it broke open and at the part fractured first there was a snowflake. I took the other end of this bar and bent it in the opposite direction and found a beautiful snowflake at the point that fractured first. I believe I can put the snowflake wherever you want it.

F. B. FOLEY.—It is probable that there is more than one kind of flake. Bar 032, on page 231, with a tensile strength of 47,000 lb. and an elastic limit of 47,000 lb., shows a defect that covers about 50 per cent. of the fractured area, see Fig. 11. This flake undoubtedly resulted from a crack and has cut the tensile strength in half. I know of one other case of a flaky bar breaking before an elastic limit had been reached, and here again the origin of the flake was probably a crack. This last-mentioned bar was found in analyzing the results obtained in 59 forgings, from which 334 bars were taken, 182 of which showed no defects, and 152 of which showed flakes. The average of the physical properties of these bars was:

	Tensile Strength, Lb. per Sq. In.	Elastic Limit, Lb. per Sq. In.	Elongation, Per Cent.	Reduction in Area, Per Cent.
182 good bars.....	101,580	68,268	19.51	37.53
152 flaky bars.....	97,282	68,120	11.59	20.48

It will be noted that the elastic limit is but slightly affected, which is what one would expect of a defect that does not become evident in breaking the bar until the elastic limit has been passed. When the elastic limit of the bar has been exceeded the defects in the bar, having no ductility, part; the crack so produced acts as a nick, thus cutting down the tensile strength, elongation, and reduction of area. If a crack exists in the bar before pulling, it will cause failure, due to the so-called "notch effect" of a crack, before an elastic limit is reached, as has been shown in the two bars mentioned. The results from broken test bars indicate strongly that, in a great majority of cases, flakes do not exist as cracks in the test bars before pulling and it seems doubtful if they ever were cracks, since the amount of reduction in forging does not seem sufficient to weld any deep seated cracks in the ingot.

Doctor Howe states that flakes are found about one-third way in from the skin of the forging. It is possible that, after the condition that causes a flake has been produced in heating the forging, the kneading of the metal by the press or hammer may correct the condition near the skin of the

ingot but still allow it to persist farther in. It is a question whether the amount of metal removed from the muzzle end of forgings is sufficient to remove the flaky metal from that portion of forgings.

In reply to Mr. Hibbard, the metal that I have had under observation was made in a basic-lined electric furnace, cast in 25-in. corrugated octagon iron molds, stripped at about 1200° to 1250° F. (648° to 677° C.), charged in a furnace and heated to 1500° F. and then removed to a forging furnace. Experiments have shown[1] that a piece of steel of a given size put in a furnace running and maintained at a given temperature will reach the furnace temperature more quickly when the temperature of the furnace is high than when it is low. So that, when the forge furnace is at a temperature of 2350° or 2400° F., the ingot heats very rapidly and causes the metal to reach a very high temperature without having been given sufficient time for the proper diffusion of ferrite.

HENRY TRAPHAGEN,* Toledo, Ohio (written discussion†).—The occurrence of flakes in steel is a perplexing problem, if we consider this condition as a disease rather than a symptom. But if considered as an exaggerated phase of a common disease, flakes will be recognized as old and well-known enemies. I have frequently found flakes in open-hearth steel, converter steel, and even electric steel; and not only in the so-called alloy steels but in dead soft 12-point carbon castings. In fact, I have some twenty flaky test bars that came from a foundry that manufactures converter steel castings.

The physical characteristics of such material are striking in their constancy, for the yield point and elastic limit are always high, the elongation is variable, and the reduction of area is nearly always negligible. Soft steels of this kind can easily be bent to an angle of 150° and the only bad symptom is the appearance of incipient cracks on the surface; when the carbon exceeds 30 points, the bending qualities are seriously affected; and when the carbon reaches the neighborhood of 50 points, the bars are generally pulled short. The elongation is peculiar in that it seems to take place along the whole length of the bar, giving a crinkly appearance such as is found in cast brass and monel metal. Of course, this elongation only takes place in the softer varieties of steel.

I have stated that flakes in steel represent exaggerated symptoms of a common disease, and that disease is the presence of higher oxides of iron. In every instance that has come to my attention, this condition is coexistent with the presence of oxides in some form. In open-hearth practice when this condition occurs, almost invariably it will be found that the metal has been loaded with ore or that tons of rusty turnings were being melted. In converter practice, one of three conditions is

[1] M. E. Leeds: Some Neglected Phenomena in the Heat Treatment of Steel. *Proc.* Amer. Soc. Testing Mat. (1915) **15**, 6.

*Metallurgist, Toledo Steel Castings Co. †Received Apr. 23, 1919.

nearly always present: bad cupola practice, the use of oxygenated pig iron, or faulty tuyeres. With the electric furnace, my experience is quite limited, but the few examples of flaky electric steel that have come to my notice were made from a charge of turnings and low-grade light scrap.

These assertions seem to be substantiated by the fact that I have repeatedly banished flakes and hackly steel by the simple expedient of throwing out all turnings and light flashings and insisting on careful melting. In one instance, shrapnel and 4.5-in. (11.4-cm.) shell forging were being repeatedly rejected because of hackly, flaky fractures and lack of ductility in the test specimen. An examination of the melting department revealed heaps of rusty turnings and equally large heaps of iron ore ready to be thrown into the furnace. An examination of the furnace slags showed a remarkable preponderance of the higher oxides of iron, ferrous oxides being outranked three and often four to one. When the turnings were kept out of the furnace and the melters were compelled to work the heat and to use much less ore, flaky and hackly fractures became rare.

A converter shop attempted to manufacture gun-mount castings under Navy B specifications. Heat after heat failed because of low reduction of area and variable elongation; but there was nothing in the ordinary chemical analyses that would indicate the source of the trouble. Microscopic examinations showed that the anneal had been sufficient and that the grain was well refined, but the microscope also showed an abundance of tiny dots embedded in the material, especially in the pearlite areas.

Thinking that forging might break up this non-metallic structure and increase the ductility of the metal, I had a series of test blocks cast, all from the same heat, approximately 4 in. (10 cm.) square and 10 in. (25 cm.) long. The blocks were all sound. One of these blocks was machined down after annealing to a standard test bar and then broken; its tensile strength was 68,000 lb. per sq. in. (4780 kg. per sq. cm.); the yield point was 47,000 lb. per sq. in. (3304 kg. per sq. cm.); the elongation, 21 per cent.; and the reduction of area, 13 per cent. The fracture was hackly with small flakes.

One of the blocks was forged down to about 2 in. (5 cm.) square, annealed, and then machined as before; there was little change in the result. The third block was forged to 1½ in. (3.8 cm.) square and the test bar cut out as before; I believe the reduction increased to 28 per cent. The fourth bar was forged out to ⅞ in. (2.16 cm.) square, a piece cut off, annealed, machined, and tested. This showed a tensile strength of 69,000 lb. per sq. in. (4850 kg. per sq. cm.); a yield point of 47,500 lb. per sq. in. (3339 kg. per sq. cm.); an elongation of 32 per cent., and a reduction of area of 52 per cent. The fracture was fine and silky. These results were confirmed by similar treatment of subsequent heats and the con-

clusion was drawn that drastic, mechanical reduction was necessary to break up this non-metallic residual structure to get satisfactory results. It was evident that the ordinary amount of forging was not sufficient, but these people were making castings and not forgings and the government needed the parts, so the problem seemed a hard one. It was evident that the trouble was due to non-metallic sonims, and I felt convinced that oxides in an emulsified form were the "niggers in the wood pile." We accordingly purchased some well-made low-phosphorus pig iron and some shell tops for scrap. The foundry sprues were kept in the yard for the first week and the cupola practice was overhauled to eliminate the possibility of oxidation due to low melting. The heats were blown with due regard to the proper angle, blast pressure, and condition of tuyeres. The molds were cleaned and kept clean and over the test-bar portion of the casting we placed a large riser, so that any sonims present would have every opportunity to agglomerate and rise. When the metal was blown, manganese was added and carefully stirred in, and then the 50 per cent. ferrosilicon was added; the heat was then held at least 10 min. to allow the churned up slag, oxides, etc. to rise to the surface.

The first heat was satisfactory, so was the second; at the end of a week the foundry was allowed to gradually work in some of the old sprues and gates and the use of the shell ends was gradually discontinued. As a result, there was not another heat lost during the life of the contract; the elongation averaged 30 per cent. and the reduction of area averaged 48 per cent. Wherever I have found rusty turnings and thin scrap, I have found woody and flaky steel; wherever open-hearth heats were loaded with ore and stopped on a rapid boil, I have found the same trouble.

I do not believe that nickel steel *per se* is especially subject to flakes, but I do believe that nickel steel made out of turnings is, like any other steel made that way, quite subject to them. We have had a veritable epidemic of flakes in nickel steel, but when we reflect that nickel was worth 80 c. per lb. and that there were thousands of tons of turnings containing an average of 3 per cent. nickel throughout the country, I cannot see the mystery of so many failures in nickel steel because of flakes.

The subject of oxides in steel and iron has had too little attention. Frenzied production in blast furnaces, coke ovens, and steel mills, at the expense of rational metallurgy, is entirely too rampant and is leaving a trail of troubles throughout the country that is destined to prove a veritable Frankenstein if not promptly checked. Thousands of tons of coke unfit for metallurgical use are being shipped. The chemical analysis is satisfactory but the structure is a joke, for the reason that the coking process has been rushed until the product resembles a carbon sponge. In the cupola and blast furnace the use of this soft, spongy

material means low melting and oxidized iron, but since such coke is sold on chemical analysis it is accepted.

Hardly a week passes in which I am not compelled to reject, in some plant or other, at least one carload of pig iron that is unfit for use. The chemical analysis is satisfactory, it always is because pig iron is sold on a chemical basis. But a chemical analysis will not account for hollow pigs, coarse open fractures, coupled with hard iron, and pigs that are literally so rotten that they break to pieces when thrown from the car. When such iron is used, the results are always the same: wild heats, skull ladles, blowy castings, and brittle iron. If the pig happens to be a steel-making iron, we have corroded linings, wild heats, cold metal, and invariably some modification of the condition we call flaky or woody structure. This form of pig iron invariably shows the peculiar compact graphite so ably described by J. E. Johnson, Jr., and it is the result of worn-out linings, spongy coke, and the use of scale, slag, or turnings, as part of the burden. The iron people profess ignorance of this condition and harp on chemical analysis, but it is a singular thing that when their attention is called to it a few times they shift this kind of material to some other shop.

Rusty turnings and flashings are all right for sash-weight foundries and shops rolling stock for reinforced concrete, but their use ought to stop right there. It is a popular fallacy that such material makes good steel in the electric furnace, but does it? The electric furnace maintains a neutral atmosphere, no doubt, but does a neutral atmosphere prevent this rust from emulsifying the steel? You may be interested in a curious phase of this subject that has recently come to my notice. On several occasions during the past year, I have been asked by tool companies to investigate a peculiar tool steel that would not harden. As usual, the chemical analysis was satisfactory; the usual bugbears, sulfur and phosphorus, were low enough to suit the most fastidious, the manganese and carbon contents were both satisfactory, and the material had been well annealed, yet the metal would crack and show soft spots, even under the most careful treatment. In every instance, I found the same condition, *i.e.*, tiny, black spots often surrounded by small white areas, and the curious part of it is that investigation showed that this steel had been made in an electric furnace from large quantities of turnings.

Flakes are not mysteries. If this flaky material is investigated from the standpoint of faulty raw materials and furnace practice, the remedy will be found. The content of higher oxides in the slags should be determined and accurate oxygen determinations made by the Walker-Patrick method; the results will prove most interesting.

When the man with the production chart gives place to the competent metallurgist, then, and not until then, will our European friends stop laughing at our flakes, transverse fissures, and quick-rusting steels, which I hold are all products of sonims, especially emulsified ferric oxides.

C. B. BRONSON,* New York, N. Y.—Mr. Wickhorst has called attention[1] to the apparent similarity between snowflakes in gun forging steel, and nuclei of interior transverse fissures in rail-heads. I have samples of the latter, which illustrate their general appearance. Those who have had an opportunity to examine both types of defects find common macro- and microstructural features in each. Furthermore, both are defects without depth, being merely surfaces of separation located within the cross-section of the rail or forging. Slag inclusions, however, are not found generally in the metal of these defective rail-heads, though they are prominent in the microstructure of flaky steel for gun forgings.

We have cut up and fractured a very large number of rail-heads that developed interior transverse fissures in service, to study the nature and cause of these defects. Pieces were split by wedging along different planes, and nuclei disclosed transversely, longitudinally, and at many other angles. Their appearance so closely resembles the snowflakes illustrated in Mr. Rawdon's paper that they may be interchangeable with them in many cases.

These two types of defects differ in one respect—their origin. Snowflakes are traced to ingot casting and cooling conditions, while nuclei in rail-heads have their inception in another manner. The conditions for the formation of the latter are cold rolling, or rapid chilling of the hot rails on the cooling beds, rendering the metal brittle with a decided loss of ductility and reduction in tensile strength, especially in the weaker and less worked interior portion of the rail-head. Complete fracture of the weakened interior metal results when the rail section is subjected to the sudden and powerful blow and overstraining action during the straightening process. Although ingot casting and cooling conditions are not held primarily responsible for the weakening of the interior metal of the rail-head, they probably exert considerable influence toward their formation and development.

H. C. BOYNTON,† Trenton, N. J.—Our practice is almost entirely with acid open-hearth steel, and we use for this purpose the purest melting materials we can buy; *i.e.*, low-phosphorus pig iron, low-phosphorus scrap, low-phosphorus iron ore; in fact, in the steel manufactured from such high-grade material as that from which our rope wire is made we rarely, if ever, find flaky spots. We make a small amount of basic open-hearth steel and occasionally, in such material, have run across a few flakes; but they appear much smaller than those described by Mr. Foley, and generally can be removed by heat treatment.

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*Of the New York Central R. R.

†Metallurgist, John A. Roebling's Sons Co.

[1] See p. 748.

I believe the rail fracture referred to by Mr. Wickhorst to be different from a "flake." We have similar breaks resulting from the "fatiguing" of a wire in a rope. If you will consider the wire in a rope running more or less continuously around a sheave; if this sheave is properly designed, that is, its diameter is not too small, the rope will last for a good many years without breaking, but if the sheave is too small the wires may begin to break in a few weeks. But with the best design of hoisting equipment possible to install, wires will finally break due to fatigue.

The accompanying illustration shows a radial fracture. An imperfection on the outside seems to have been the nucleus or starting point of the rays in the break. Such radial fractures nearly always start at some slight, almost microscopic, imperfection in the wire, which imperfection may be a little nick or a non-metallic impurity or "sonim." But in nearly every case where premature fatigue takes place in a wire the nucleus of the radial lines has been a "sonim." I feel quite certain that such a fatigue break differs from a flake as described by Mr. Foley. In the fracture in question there may have been just such a sonim in the head of Mr. Wickhorst's rail and years of flexing at this point gradually developed a fatigue break. One has only to watch a train at any crossing to see the rails bend up and down, very well illustrating this fatiguing action.

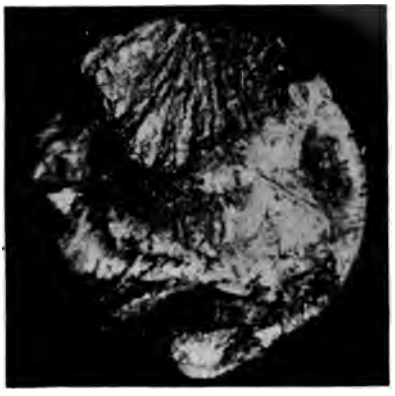

FIG. 5.—BREAK OF A "FATIGUED" WIRE. FRACTURE DEVELOPED FROM AN IMPERFECTION SEEN AT EXTREME TOP OF PICTURE, AND PROGRESSED RADIALLY FROM THIS POINT. × 15.

THE CHAIRMAN (H. M. HOWE, Bedford Hills, N. Y.).—Will Doctor Boynton tell us whether he refrains from the use of alloy steels?

H. C. BOYNTON.—We always use carbon steels wherever possible, for in making rope wire we want long pieces to avoid splices, because every splice is a source of weakness. With straight carbon steel we manufacture single bundles of rope wire weighing 400 lb. and over and having a tensile strength of over 400,000 lb. per sq. in. (28,120 kg. per sq. cm.) so that we seldom need to employ alloy steels. Our ingots average about 10 or 12 in. (25 or 30 cm.) square in cross-section, and in these sizes we seldom run into any flakiness.

A. A. STEVENSON.—Have we exhausted the possibilities of carbon

steel? Personally, I think not. My own feeling is, and I think it is the feeling of some others, that primarily flakes are due to original ingot condition—a condition that might be avoided had we a more accurate way of determining temperatures. I think one of the greatest aids we could have in our metallurgical operations today is some practical instrument for measuring the temperatures of the bath in an open-hearth furnace.

CHAIRMAN HOWE.—The combination of circular and radial lines shown is interesting. The development of a fracture through glass is the same. Mr. Stevenson says that the trouble is thermal and the remedy depends on the development of a pyrometer that will register the same thing under the same conditions for all kinds of steel. Doctor Burgess, of the Bureau of Mines, has been studying that subject and has made such progress that we feel that we shall have within a reasonable time a pyrometer for use with open-hearth furnaces.

H. C. BOYNTON.—We know we cannot get along without alloy steels but we get better results by not using alloy steels when plain carbon steel will answer the purpose.

CHAIRMAN HOWE.—The Englishman always speaks about cogging and I wondered for a long time what he meant. When a man starts to draw out an ingot he cogs it like the cogs in a wheel. As a result the ingot is given an initial spread sideways; that is the reason why they cog crucible steel ingots satisfactorily with transverse forging, so that it is a question whether it is not advisable, where transverse tests are to be made, to begin with a spread and then proceed to draw out. We are told that better transverse test figures will be obtained if you start with a transverse spread.

J. E. JOHNSON, JR.—I understand that the consensus of opinion favors the suggestion that flakes were caused by a crack in the ingot. Why should a crack form in homogeneous molten metal and, especially, why should such a crack form in the middle third of the ingot? It has occurred to me that these flakes may be caused by recalescence of the steel producing an expansion; and especially that recalescence occurring near the blue-heat stage of the steel. The fact that they are more numerous in some steels than in others may be caused by the possible circumstance that the tender stage of one steel is nearer the recalescence point than it is in others. Thus, for example, in carbon steel, the tender stage and the temperature of recalescence may be so far apart that no flakes are caused.

BOYD DUDLEY, JR.,* Watervliet, N. Y.—The papers presented at this meeting on the subject of flakes in steel and the discussion of them

*Captain, Ordnance Dept., U. S. A.; Superintendent, Heavy Field Cannon and Metallurgical Departments, Watervliet Arsenal.

throw much light on the nature of these defects, and on their causes and the conditions that control their origin. We seem fairly well agreed as to what flakes are and, although we are somewhat less in harmony as to their cause, we should be able to improve materially the quality of our steel by correcting our practice in the several ways that have been suggested, on the assumption that any or all of the theories on the origin and cause of flakes may be correct.

I believe, however, that in discussing the nature and causes of flakes we have lost sight of the question of their practical effect upon the physical properties of the steel as a whole. Of course we all agree that flakes are undesirable, that their formation should be prevented by improved casting and forging practice, and that the physical properties of the steel are impaired by their presence. But the extent to which they render the steel unsafe for use and the means of determining the degree to which its safety is impaired by their presence are things that have not been touched upon.

Since flakes frequently are discontinuities in a steel of otherwise normal structure, their practical effect upon its physical properties depends on their size, number, and arrangement in the metal. Fairly numerous small flakes, when uniformly distributed, may exert but little detrimental influence on the physical properties of the steel.

Our present active interest in flakes has resulted from the difficulties recently encountered by steel manufacturers in the production of forgings for ordnance. Both the Army and Navy specifications require the use of physical test specimens of 0.505 in. (12.8 mm.) diameter, and the failure of such specimens to meet the specifications results in rejection of the forgings represented by them. If such a failure is directly caused by the presence of flakes and not by unfavorable physical properties of the sound parts of the metal, the failure results because the specimen happened to contain flakes of such a size in relation to its cross-section and in such a position in relation to its axis as to impair the physical properties of the specimen as a whole. Is such steel actually unfit for use? Are we measuring its actual and practical strength from the standpoint of use in cannon by such tests? I believe it is self-evident that in many cases of flaky steel better test results may be expected from the use of specimens of larger cross-section, and the practical strength of cannon forgings made from such steel could be determined only by specimens approximately equal in thickness to that of the finished member.

Last summer, the Watervliet Arsenal was called upon to produce, in a very short time, a number of experimental field guns for special caliber. The forgings for these guns were received at the Arsenal in the rough-machined, heat-treated, tested, and accepted condition. During the course of operations it became necessary, because of a loss of identifica-

tion marks from two of the jackets, to subject them to further physical tests. Only enough metal remained to allow the use of specimens of 0.357 in. (9 mm.) diameter. The first pieces that were pulled failed to pass the usual specification due to the presence of flakes, which were small in size but sufficiently large to produce a very considerable lowering of the physical properties of the specimens. This led to the retesting of six jackets. Three specimens 0.357 in. in diameter could be obtained from each. Practically all of these specimens failed to pass the specifications because of the presence of flakes, although the test records of these jackets obtained by the inspector at the forging plant and based upon the usual 0.505 in. diameter specimens were perfectly clear.

The guns in question were of special and very light design and the usual factor of safety was not present. In fact, the factor of safety based upon the minimum specified elastic limit of the steel and the calculated service pressure was only slightly greater than unity The guns were so urgently required that it was decided to use the jackets with their unfavorable tests rather than to wait for others. Several of these guns have been successfully proof-fired without abnormal increases in bore diameters and, apparently, the steel was as good as the 0.505 in. diameter specimens indicated it to be. Certainly it was better than the 0.357 in. specimens would lead us to believe. It is doubtless a fact that many flaky cannon forgings are actually better from the standpoint of safety in service than test results with a 0.505 in. specimen would indicate.

I have not made these statements and presented these facts with the idea of belittling the evil effects of flakes; unquestionably our cannon forgings will be better if flakes are entirely absent. But I wish to emphasize the fact that the methods of sampling and testing now prescribed may not, and probably do not, serve to measure accurately the actual and practical strength of the steel in such forgings. In defense of these methods it probably can be said that in such cases the error in measurement introduced is on the safe side. But in time of war, when every ton of steel means much, we should take all possible steps consistent with safety and good service to prevent the rejection of material that actually is suitable for use.

F. B. Foley.—We have not reached the point where we can say definitely that a flake is or is not a crack. It certainly appears that, if a flake is evidence of a crack, it does not exist as such in the finished heat-treated forging nor in the tests taken from the finished forging until, in pulling the test, the elastic limit has been exceeded. A crack may give an appearance, to the fracture of a test bar, which is very similar to that of a flake, in that the defect may be crystalline in the midst of metal of a fibrous or silky character, but I believe that a fracture of this kind can be distinguished from the kind of fracture I know as flaky, and also that the tensile strength and elastic limit of the material

will be affected to a much greater degree than they are in the case of the ordinary flaky metal in which the elastic limit is practically not affected and the tensile strength is reduced about 4000 lb. per sq. in. (281 kg. per sq. cm.). This reduction in tensile strength is to be expected since the defect becomes apparent only after the elastic limit has been exceeded. Were the defect a crack in the beginning, it would in all probability lead to failure long before the tensile strength of the good metal had been reached, depending of course on the size of the crack. We should expect at any rate a great number of rejections due to very low tensile strength, but as a matter of fact failures have not been due, in many cases, to this but to low ductility alone. It is suggested that flakes are due to cracks that have been welded under the press, which may be so if we allow that the reduction is sufficient to accomplish this.

On the supposition that flakes are produced by overheating the ingot preparatory to forging, I gave a Leeds & Northrup optical pyrometer to one of the men at a plant where flakes were a source of great trouble and instructed him not to permit any forging to be heated to above 2200° F. (1204° C.). About three or four days later, the operator was transferred elsewhere in the plant and the use of the pyrometer was discontinued. About the time when these forgings were due for testing twelve out of fifteen were accepted on first submission. It is not possible to draw definite conclusions from this performance because the plant failed to keep records by which these forgings could be definitely identified as those that had been under pyrometric control during heating for forging, but it is noteworthy that no such record of freedom from flakes was approached either before or after the pyrometer was used.

I fail to see why defects that are the result of cracks should be confused with true flakes. The fractures are quite different in appearance and recently I was able to say immediately from the appearance of the fractures of two different tests that the defects in the test bars were the result of cracks that existed in the test bars before pulling and, furthermore, I was able to make a qualitative estimate of the physical properties. The records on the books checked my estimate, one bar broke with a tensile strength of about 40,000 lb. and the other gave about 65,000 lb. Fig. 11 (p. 218) is a typical fracture resulting from a crack that existed before the test was pulled. It will be noted that this bar gave a tensile strength of but 47,000 pounds.

H. D. Hibbard.—I am not able to form any opinion about the steels referred to because their history has not been told. Not much progress can be made on investigations that begin with the finished steel; it is necessary to know the conditions and treatment of manufacture. Presumably the defective steels have not been well made, due, in part no doubt, to war pressure, which ever demanded more steel.

Two years ago I saw in Sweden perhaps the finest open-hearth steel in the world. The superintendent said that some of the melters could regularly bring out their heats within 10° C. of the proper temperature; a statement which, at that time, seemed exaggerated but which, later, I came to believe to be true. The steel, which was for flying machines, was cast into solid ingots that rolled beautifully in the blooming mill without a crack. A skillful melter, with a suitable plant and materials with which he is familiar, can make heat after heat of good steel with rarely a miss.

I have little faith in the use of pyrometers to control casting temperature, but the melter has a number of things that aid him in such control. He is familiar with the general intensity of light from his furnace, and the relative intensities of the flame, back wall, surface of the slag, and the wakes of the bubbles rising from the metal beneath. Then he gets great aid from taking furnace tests or samples. He notes the effect on the rod with which the bath is stirred, the appearance of the metal in the test cup, the time it remains molten in the cup, the skull remaining, if any, and the rate of solidification in the test mold. These, bearing in mind the degree of oxidation of the metal (which affects its fusibility) as evidenced by the rate of boiling and basicity of the slag, give him an accurate conception of the state of his charge and tell him what steps to take to correct it if it is not as desired.

I took Caspersson's table of 1882 relating to Bessemer steel casting temperatures, assumed the extreme temperatures and interpolated the others with the showing that 10° either side of the ideal for a given ingot structure would give inferior ingots and I am willing to grant the claim of my Swedish friend. Caspersson, of course, did not know the absolute temperatures of his steels. His temperatures were wholly relative.

FEDERICO GIOLITTI,* Torino, Italy.—In Italy, we are very much interested in the defects known as flakes and woody structure, especially in the Ansaldo Works where about two-thirds of the guns used by the Italian army are made. Though none of the gun mountings manufactured at our works have been rejected on account of these defects, they have been known to us for some time, but through experimental work we have been able to eliminate them almost entirely.

We believe flakes originated in tender alloy steels as intercrystalline cracks, probably intensified by inclusions and segregations. They can consequently be controlled by careful steel-furnace practice, casting, soaking, reheating, forging, and heat treating. Contrary to Doctor Howe's recommended practice, we pour very hot. Our forging ingots are fluted octagons, and what flakes occur have no habitual location or direction. The large dentritic crystals formed in hot-poured ingots re-

* General Manager of Steel Works, Giov. Ansaldo & Co.

quire care and skill in forging; but when the work is rightly done, such steel produces superior results. There is a saying in our plant that when the outer part of the hot-poured ingots has a tendency to crack under forging, it is a sign of a good heat and no rejections will occur.

Chromium steel properly made of pure materials under reducing conditions and containing no oxide will never crack. You can pour it into an ingot of any cross-section and forge it as you like, always providing the work is done within the correct temperature limits and with skill, and flakes will not appear. On the other hand, a bath of metal made from correct raw materials will excessively develop woody structure if only a few kilograms of scale are thrown into the furnace, even a long time before tapping. If rusted scrap or turnings are melted the chances for rejection are greatly increased. Again, we have found trouble as soon as the only raw materials available were of inferior grade. Such experiences as these convince us that flakes and woody structure, which two defects are near relatives, occur only in steel containing highly oxidized inclusions.

It seems to me that the presence of ferrite near a slag inclusion is due rather to decarbonization caused in a manner somewhat like the following: There exists in slowly cooled clean steel (normalized steel) a considerable quantity of gas in solution, such that a substantial equilibrium exists between the system $Fe:Fe_3C:CO:CO_2$ according to the following equation

$$3Fe + 2CO = Fe_3C + CO_2.$$

The gases CO and CO_2 are occluded or dissolved in the metal in such a manner that no cavities are found. Similar equilibrium conditions of an oxidized slag may readily change with changing temperature in such a way that an excess of CO_2 is transferred into the metal unbalancing the equation given above and shifting it to the left, actually producing a zone of decarbonized iron in that vicinity, somewhat supersaturated with CO gas as well. In this way a spot of ferrite appears about an oxidized nucleus. On working, this spot flattens out into a thin sheet perfectly continuous with the metal behind it, and only separated at perhaps a minute point in the very center, but the flake is not in the real sense an internal crack. All these matters have been discussed by me completely in different papers published, in 1914, in *Zeischrif fur Metallographie* and, in 1916–17, in *Metallurgia Italiana*. If my gas theory is true, it should be easy to cure woody structure by heat treatment. This has been done in the Ansaldo works when the steel was not actually overloaded with highly oxidized slag If not excessive, woody structure can be diminished by careful reduction at temperatures of 1200° to 800° C. The exact explanation of this fact is not clear, except perhaps that forging a low heat destroys Widmannstättian structure and with it any ferrite cleavages.

Metallography of Rifle-barrel Steel*

BY G. F. BUTTERWORTH JR.,† A. B., LL. B., NEW YORK, N. Y.

(Chicago Meeting, September, 1919)

THE metallographic structures most frequently encountered in rifle barrels, and which are illustrated by the accompanying photomicrographs, fall naturally into two groups, distinguished by the method used to produce in the stock a physical condition having the requisite properties. The first group consists of "rolled" barrels; that is, barrels subjected to hot working by rolling in or near the critical range. In the second group, the stock, which is smaller in diameter than the first, is upset to form the butt end and is then heat treated by giving it a quench and a draw. These barrels will be referred to as "heat-treated" barrels.

The structure of the rolled barrels resembles closely that of the same steel after annealing. There is the same network of excess ferrite outlining the grain boundaries, but the grains themselves are composed of sorbite rather than pearlite. The grain size is closely related to the rolling temperature. The critical-point curves of this grade of steel, which is approximately 0.50 to 0.60 per cent. of carbon and 1.00 to 1.30 per cent. of manganese, show a single very pronounced point between 1300° and 1350° F. (704° and 732° C.). Barrels rolled within this temperature range give an exceedingly fine grain, shown in Fig. 1. In fact, the grains may be so fine that, at a low magnification, they may be confused with the heat-treated structure shown in Fig. 6. A magnification of 500 diameters, however, will always resolve a rolled barrel into the characteristic structure shown in Fig. 2. As the temperature is increased, the grains are found to be larger. If, on the other hand, the rolling temperature is below the critical range, the structure previous to rolling will not be obliterated, the only effect of rolling being to elongate the coarse sorbite grains in the direction of rolling, Fig. 3. This distortion is greatest at the muzzle.

Heat-treated barrels are quenched in oil from above the critical range, which should give a martensitic structure, Fig. 4, but the presence of some troostite in the martensite is frequently noted when the quench has not been sufficiently drastic, Fig. 5. The structure brought about by the subsequent draw is not so striking, though quite as typical. It is sorbitic or sorbito-pearlitic, and under a low magnification appears almost homogeneous, as in the case of fine grains of pepper and salt well mixed together, Fig. 6. This structure is substantially the same for all

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† Metallographer, U. S. Armory, Springfield, Mass.

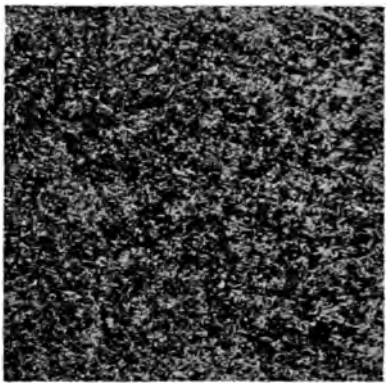

FIG. 1.

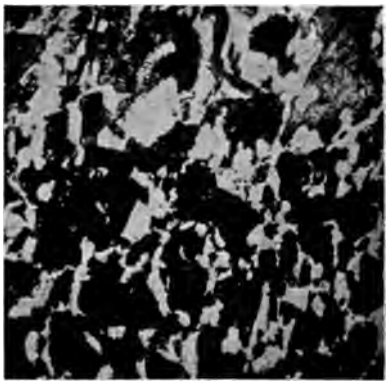

FIG. 2.

FIG. 1.—BARREL ROLLED AT 1350° F. (732° C.). × 100.
FIG. 2.—BARREL ROLLED AT 1350° F. (732° C.). × 500.

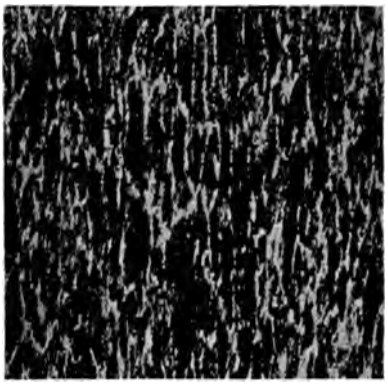

FIG. 3.

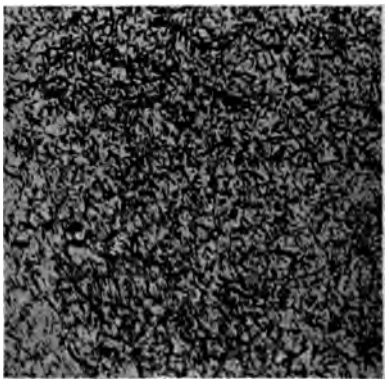

FIG. 4.

FIG. 3.—BARREL ROLLED BELOW CRITICAL RANGE. MUZZLE END. × 100.
FIG. 4.—BARREL GIVEN DRASTIC QUENCH FROM ABOVE CRITICAL RANGE. × 100.

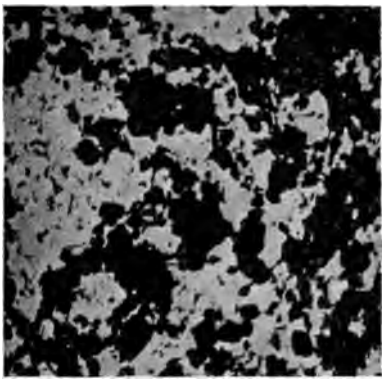

FIG. 5.

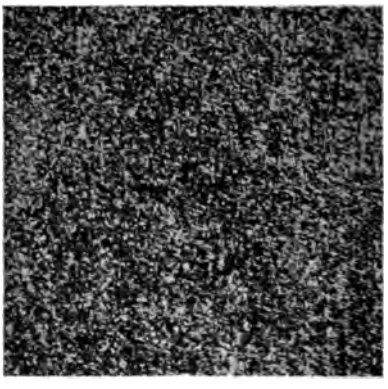

FIG. 6.

FIG. 5.—BARREL GIVEN LESS DRASTIC QUENCH. × 100.
FIG. 6.—QUENCHED FROM 1500° F. (816° C.) AND DRAWN AT 1200° F. (648° C.) × 100.

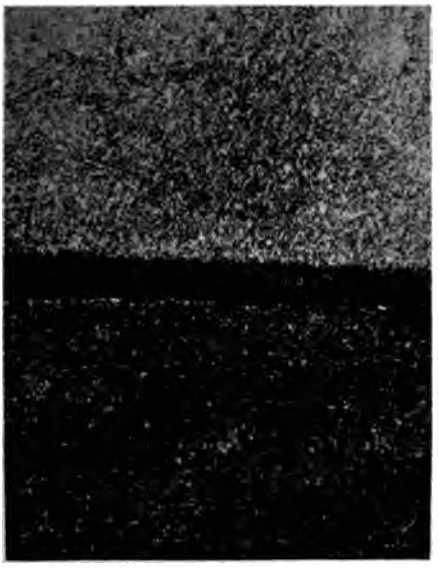

FIG. 7.

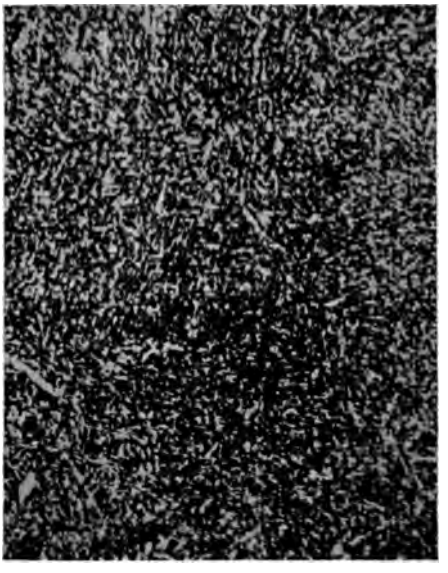

FIG. 8.

FIG. 7.—BARRELS QUENCHED FROM 1500° F. (816° C.) AND DRAWN AT 800° F. (427° C.) (DARK) AND 1200° F. (648° C.) (LIGHT). × 100.
FIG. 8.—BARREL QUENCHED FROM 1500° F. AND DRAWN AT 1200° F. × 500.

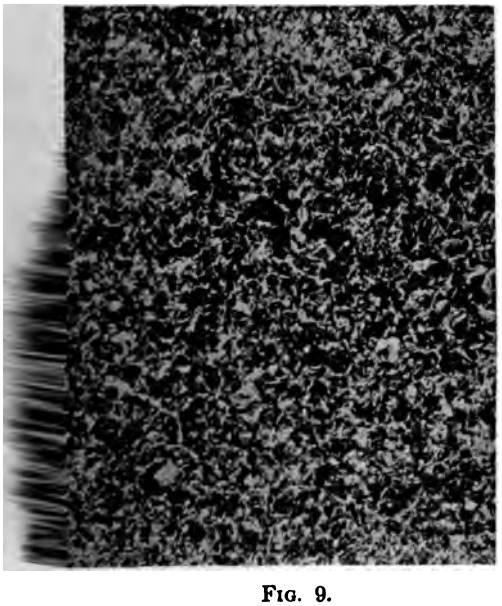

FIG. 9.

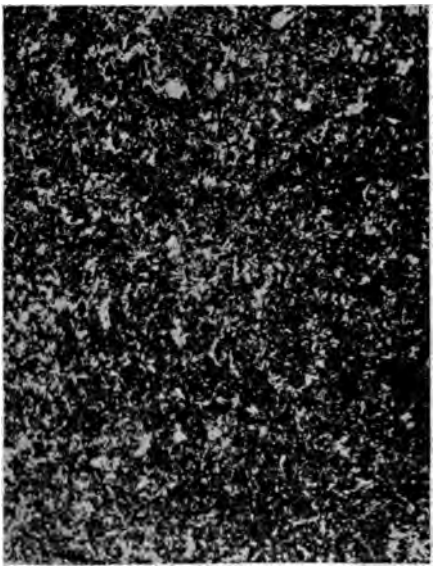

FIG. 10.

FIG. 9.—BARREL GIVEN INEFFECTIVE QUENCH. STRUCTURE NEAR BORE. × 100.
FIG. 10.—SAME. STRUCTURE NEAR OUTSIDE. × 100.

FIG. 11.

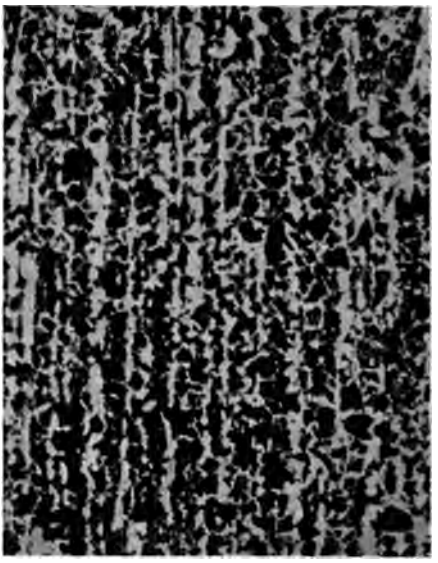

FIG. 12.

FIG. 11.—STEEL BURNED IN UPSETTING. × 35.
FIG. 12.—BANDED STRUCTURE. × 100.

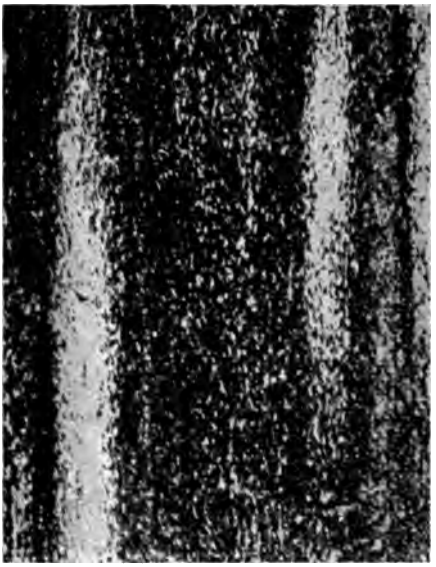

FIG. 13.

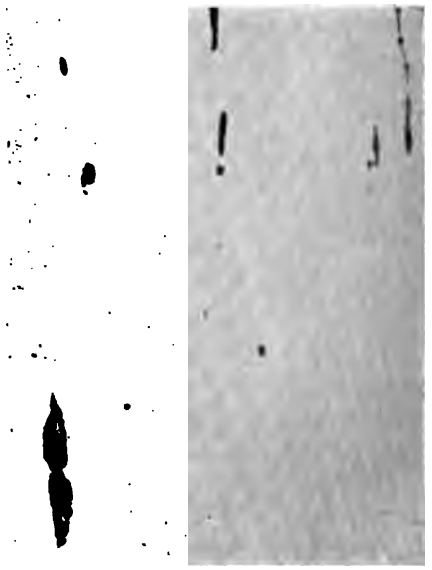

FIG. 14.

FIG. 13.—PHOSPHORUS SEGREGATION ETCHED WITH STEAD'S REAGENT. × 100.
FIG. 14.—STRINGY SLAG IN LONGITUDINAL SECTION. × 100.

drawing temperatures from 800° to 1300° F. 427° to 704° C., 1200° 648°, being the temperature most frequently used. In fact, it is almost impossible to determine, even roughly, by microscopic examination of its structure, the temperature at which a barrel has been drawn, but it is possible to estimate it within 100° F. by a macroscopic examination. It has been found that the color of the specimen varies with the drawing temperature, the color becoming lighter as the temperature increases. This is illustrated by Fig. 7, which shows specimens drawn at 800° and 1200° F. (427° to 648° C. respectively, photographed side by side. The color contrast is greatest after etching the specimen for 45 sec. in a 5 per cent. solution of picric acid in alcohol, or for 6 sec. in a 5 per cent. solution of nitric acid in alcohol. The use of the former is recommended, as it will reduce the effect of the unavoidable differences of time in etching the specimens to be compared. Experiments so far made have shown no variation in color after drawing, due to differences in the original quenching temperature, and the relation of this factor remains to be determined. Under the higher magnification, the heat-treated structure may be resolved into fine light-color steels in a dark matrix, as shown in Fig. 8.

The following are the most frequent microscopic evidences of defective heat treatment: If the barrel is quenched before it has been soaked through at a proper temperature, it will show large amounts of excess ferrite collected generally near the grain boundaries. These white areas are more pronounced near the bore than near the outside of the barrel, as appears from a comparison of Figs. 9 and 10. The former shows the structure near the bore, and the latter near the outside of the same barrel. If the drawing temperature is too low, the specimen will appear very dark in a macroscopic comparison with one drawn at the proper temperature. If the temperature has been too high, but below the critical range, the specimen will be lighter than the standard; and if drawn to or above the critical range, the network of excess ferrite will again appear.

Certain other defects, not due to heat treatment but which appear in the finished product, are: Barrels are sometimes burned in upsetting the butt end, with the result that the grains, enormously enlarged, are forced apart, leaving dark cracks between, as shown in Fig. 11. A more frequent defect, occurring in both rolled and upset barrels, is the presence of very marked bands in the steel, which the normal heat treatment is unable to obliterate, Fig. 12. Sometimes this structure may be traced directly to a segregation of phosphorus by etching the specimen with Stead's reagent, Fig. 13. The most frequent defect is the presence of slag, appearing in the longitudinal section as strings, Fig. 14. There is always some of this present, but, unless it occurs in abnormal amount or in seams, it appears to have no injurious effect.

Erosion Tests of Rifle Barrels *

BY A. E. BELLIS, † S. B., SPRINGFIELD, MASS.

(Chicago Meeting, September, 1919)

THERE is a wide difference of opinion among rifle experts in the matter of barrel steel, and the relative importance to the life of the barrel of the steel's composition, heat treatment, structure, and physical properties. In order to promote discussion, which it is hoped may lead to the discovery of the source of some of the traditional beliefs on the subject, the results of some tests are presented here, with data on heat treating, physical tests, and erosion. The detailed metallographic study of this steel is given in another paper.[1] The conclusions from these tests should be capable of wider application than their direct bearing on the barrel-steel problem.

The Government specifications for barrel steel for model 1903 rifle (Springfield) call for carbon 0.50 to 0.60 per cent., manganese 1.00 to 1.29 per cent., silicon under 0.25 per cent., sulfur under 0.06 per cent., phosphorus under 0.08 per cent. The minimum physical requirements are: Tensile strength, 110,000 lb. per sq. in. (7733 kg. per sq. cm.); yield point, 75,000 lb. per sq. in (5472 kg. per sq. cm.); elongation, 20 per cent.; reduction of area, 45 per cent. Material fulfilling these requirements gives practically no failures when proof-fired, with charge of 70,000 lb. per sq. in. pressure, which is 40 per cent. in excess of the service charge. The turning and drilling operations show up any seams or laps such as accidentally occur even in the best heats of open-hearth steel, so that faulty stock is practically all eliminated before proof-firing.

These physical requirements are met either by rolling billets at a proper temperature down to the form of tapered barrel blanks or by heat treating bars that have been upset to enlarge the breech section. The transformation temperature of this steel, on heating, is around 1350° F. (732° C.), varying somewhat with the composition and the rate of heating. The actual treatment of the barrel blanks, as carried out by different manufacturers, gives four kinds of barrels, which will be called *A*, *B*, *C*, and *D*, for convenience.

A barrels result from heat treating upset blanks by quenching at

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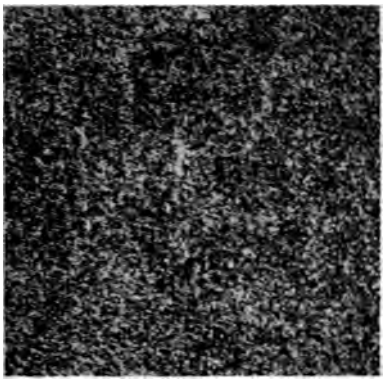
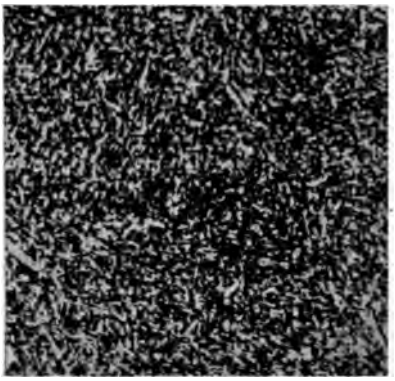
† Major, Ordnance Dept., U. S. A.

[1] See p. 287.



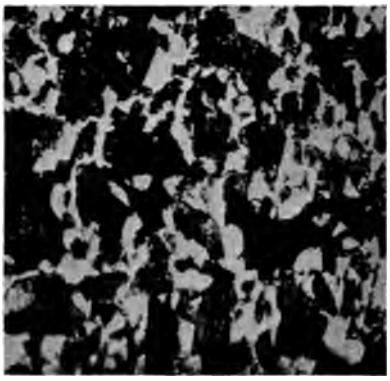
× 100.



× 500.

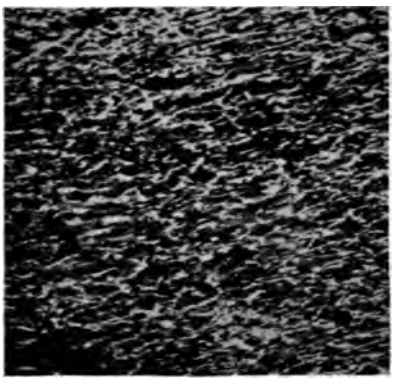
FIG. 1.—"A" BARREL.

× 100.



× 500.

FIG. 2.—"B" BARREL.



× 100.

× 500.

FIG. 3.—"C" BARREL.

1500° F. (816° C.) in oil and drawing at 1200° F. (648° C.), which is below the critical temperature, for 2 hr. *B* barrels are made by rolling the steel at or above 1350° F. (732° C.), the critical temperature of the steel. *C* barrels are made by rolling at 1200° to 1250° (648° to 677° C.) or below the critical range. *D* barrels are made by quenching the billets from 1500° F. (816° C.) in oil and then rolling at 1250° (677° C.). The resulting structure and physical properties of the *D* barrels are the same as *A* barrels. This treatment was developed in order to raise the physical properties of *B* and *C* barrels that did not fulfill physical requirements.

A and *D* barrels give a structure of homogeneous sorbite, as shown in Fig. 1. *B* barrels give homogeneous sorbite but on higher magnification (× 500) small rounded particles of unabsorbed free ferrite are observed, as shown in Fig. 2. *C* barrels generally show evidence of cold work, especially at the muzzle end, in a longitudinal section, as shown in Fig. 3. Large masses of free ferrite are characteristic of this kind of barrel.

The real test of the quality of a barrel is the number of rounds that can be fired without loss of accuracy from excessive erosion. Erosion can be directly measured by the increase in the size of the bore after firing. The extent to which the nature of the steel is a factor in erosion is a much discussed question, one on which every rifle expert has "dope," but one in which there are few published data based on facts. The importance of accuracy and barrel life, especially in machine-gun barrage work, make it essential that all the metallurgical factors of the problem be appreciated. The subject was brought forcibly to the writer's attention by the metallographic examination of a short-lived barrel showing structure *C*. A barrel giving much longer life, under the same conditions of firing, gave structure *A*. Most of the rifle experts consulted believed that other factors than the structure of the steel were the primary cause of the excessive erosion. It is well known that the dimensions of the bore and chamber, the analysis of the steel, the conditions of firing, particularly with respect to temperature, and the direct temperature effect of the powder, are important factors of erosion.

The following test was therefore planned, in which all these factors were kept constant and careful records kept of the treatment, structure and other metallurgical factors: Three barrel blanks were prepared from the same bar of steel and treated differently so as to give typical *A*, *B*, and *C* barrels, see Fig. 4. The *A* barrel was heat treated by quenching from 1500° F., and drawing for 2 hr. at 1200° F. (648° C.). The *B* barrel was rolled at 1350° F. (732° C.); and the *C* barrel was rolled at 1200° F. (648° C.). Physical tests made from blanks given parallel treatment gave the results shown in Table 1.

In the physical tests, the specimen was taken from as near the large end of the barrel as possible. The steel analyzed carbon 0.48 per cent.,

TABLE 1.—*Results of Physical Tests on A, B, and C Barrels*

Yield Point, Pounds Per Square Inch			Tensile Strength, Pounds Per Square Inch			Elongation, Per Cent.			Reduction of Area, Per Cent.		
Barrel *A*	Barrel *B*	Barrel *C*	Barrel *A*	Barrel *B*	Barrel *C*	Barrel *A*	Barrel *B*	Barrel *C*	Barrel *A*	Barrel *B*	Barrel *C*
80,000	71,700	77,000	106,150	116,050	110,750	23	25	16.5	59	53	44.9
81,450	70,300	86,900	106,400	114,750	124,550	22	22	20.1	60	53	47.1
80,600	71,150	83,550	106,150	116,150	120,950	24	24	19.6	60	55	47.8
81,050	70,550	82,650	105,650	114,950	122,800	23	23	18.5	59	53	39.4
84,500	70,750		109,450	116,050		23	22		56	54	

manganese 0.98, per cent., sulfur 0.050 per cent., phosphorus 0.060 per cent. As the minimum amounts of carbon and manganese were present,

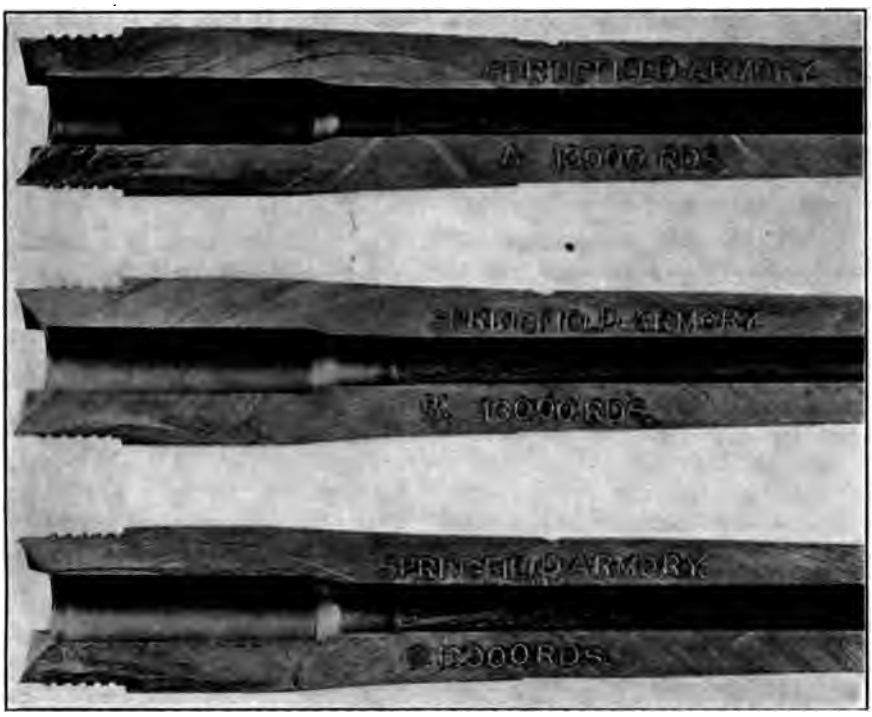


FIG. 4.

conclusions reached with this material would be emphasized with higher amounts of these elements.

These barrels were then submitted to a 13,000-round firing test with measurements for accuracy and erosion. The accuracy tests consisted of two targets of ten rounds each at 500 yd. with muzzle rest and were made after each 2000 rounds from 4000 to 10,000 and then every 1000 to 13,000. The firing was at the rate of 10 to 12 rounds per

minute per rifle, cooling the rifles with water after each 100 rounds. The same lot of ammunition was used throughout the test. The results are summarized in Table 2.

TABLE 2.—*Erosion Table*
(Unit = 0.0001 in.)

	Barrel *A*		Barrel *B*		Barrel *C*	
	Lands	Grooves	Lands	Grooves	Lands	Grooves
Breech	85	35	90	45	90	78
Muzzle	18	4	20	4	29	6

The relative erosion is shown by the increase of diameter of the lands at the muzzle, these being 0.0018, 0.0020, and 0.0029 in. for barrel *A*, *B*, and *C*. Barrel *B* gave a good target (7.3 in. mean radius at 500 yd.) after the 13,000 rounds. Barrel *C* shot off the target after 11,000 rounds. Barrel *A* shot off the target after 12,000 rounds. Barrel *A* had a mechanical defect, viz., a slight pocket about 10 in. from the muzzle, which was discovered when the barrel was sawed lengthwise to expose the bore.

This experiment shows that the homogeneous structure of a heat-treated barrel (*A* or *D*) offers the best resistance to erosion. The network of free ferrite of the *C* barrel apparently offers easily eroded channels to the washing-out action of the hot gases and metal. The appearance of the section of the barrel confirms this explanation. The uniform structure of the *B* barrel has better erosion-resisting properties than the *C* barrel, but not as good as the heat-treated barrel.

Shimer Case-hardening Process

BY JOSEPH W. RICHARDS,* SO. BETHLEHEM, PA.

(New York Meeting, February, 1919)

THERE are two essentially different types of case-hardening processes; that using a dry mixture in which the object to be case-hardened is packed and kept for the necessary time at the necessary temperature, and the "liquid" process, employing a bath of fused salts into which the object is immersed and which by immediate contact case-hardens the surface of the article. The Shimer process belongs to the second class, and was a war emergency invention.

Of the liquid melts used as baths, the most effective and most frequently used are melted potassium cyanide and melted sodium cyanide. These may be used either pure or mixed with salts that reduce the melting point and the percentage of cyanide present. The use of these baths since 1914 has met many commercial difficulties because of the high price and frequent commercial scarcity of the cyanide salts; at present sodium cyanide has practically entirely displaced potassium cyanide. Another trouble is the danger to the workmen handling the cyanide salts, which are extremely poisonous, and the annoying and poisonous vapors or gases given off in the workshop, unless an effective system of hoods and ventilation is provided, so as to prevent the gases from the baths mixing with the air of the room.

Porter W. Shimer, of Easton, Pa., has invented a substitute[1] for the bath of melted cyanides which case-hardens with equal or greater facility and effectiveness, gives off no poisonous vapors, and costs for chemicals materially less than the cyanide costs in the previously used baths. The process has been in use over a year in a large American works and the following statement embodies the results of practical experience in the use and operation of the process.

The Shimer liquid or melted bath consists of a mixture of easily fusible salts that do not possess case-hardening properties, into which is immersed fresh calcium cyanamid, which imparts to the bath case-hardening properties. The composition of the non-case-hardening salts appears rather immaterial. Good results have been obtained by using a mixture of sodium chloride, calcium chloride, and barium chloride in equal proportions by weight; also a mixture of one part sodium chloride

* Professor of Metallurgy, Lehigh University.

[1] U. S. Patents 1279457 and 1279458, Sept. 17, 1918.

to one part calcium chloride Potassium chloride can replace the sodium chloride where the question of cost is not material, producing a very liquid bath when equal chemical parts of the two salts are used; that is, 58.5 parts of sodium chloride to 75.5 parts of potassium chloride.

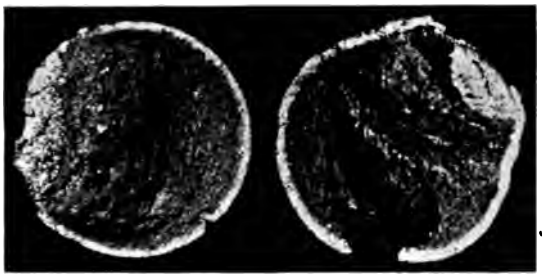


FIG. 1. FIG. 2.

FIG. 1.—FRACTURE. 1 HR. IN CYANIDE BATH. 1500° F. QUENCHED IN WATER. × 5.
FIG. 2.—FRACTURE. 1 HR. IN CYANAMID BATH. 1500° F. QUENCHED IN WATER. × 5.

Alkaline carbonates or alkaline hydroxides have also been added to the bath material with advantage in some special cases.

The mixture of non-case-hardening salts is melted in an iron or steel pot suitable for case-hardening operations, and the calcium cyanamid is brought into contact with it, which may be accomplished in several

FIG. 3. FIG. 4.

FIG. 3.—1 HR. IN CYANIDE BATH. 1500° F. QUENCHED IN WATER. PICRIC ACID. × 40.
FIG. 4.—1 HR. IN CYANAMID BATH. 1500° F. QUENCHED IN WATER. PICRIC ACID. × 40.

ways. One very effective method is to place small lumps of the cyanamid in an iron basket, which is sunk to the bottom of the case-hardening pot. A lively evolution of gas soon takes place, the exact composition of which has not yet been fully determined. The bath quickly acquires case-hardening properties, which last as long as the evolution of gas

continues. What the exact chemical reaction of the cyanamid upon the non-case-hardening salts is, to produce a melt that has excellent case-hardening properties, has not yet been determined; it would need thorough and arduous chemical investigation to precisely illuminate the rationale of the operation. The fact remains, however, that contact of the cyanamid with the other salts imparts to the liquid bath active case-hardening properties.

The active evolution of gas is a valuable feature of the process; it keeps the bath in circulation and thereby equalizes its temperature, it accelerates the heating and case-hardening of objects immersed in the bath, and finally it is an important indication to the workman that the bath is in working order, with active case-hardening properties.

In practice, the calcium cyanamid is immersed in the bath of melted salts and as soon as a lively evolution of gas is shown, the dipping in of articles and their case-hardening can be proceeded with. If the evolution of gas becomes too active, the cyanamid may be removed and case-hardening can be proceeded with for some time after this removal. When the case-hardening power of the bath decreases, the cyanamid may be re-immersed and the operation continued as before. If the cyanamid is in large fresh pieces and the evolution of gas is not too violent, the cyanamid may be left permanently in the bath until it has lost its power of imparting case-hardening properties to the melt, as is shown by the diminution of the evolution of gas. On removing this apparently exhausted material, the larger pieces may be broken, thus exposing fresh surfaces, and the material will be found to still retain active properties when re-immersed in the bath.

It has been found that the cyanamid is best used in lumps varying from the size of a walnut to the size of an egg. It should be in the fresh condition as it is taken from the furnace; that is, it should be kept hermetically sealed until used. If fine powder is put into the bath, it is difficult to keep the powder immersed, and the frothing is voluminous and troublesome. If the cyanamid has been exposed to air, absorbing moisture and becoming oxidized, it causes violent frothing when immersed in the bath, which continues an inconveniently long time. If only such cyanamid is available, it may be mixed with pulverized hard pitch or with tar, and the mass coked at a red heat; this eliminates absorbed moisture and changes the structure from powder to a porous coke. Such porous coke is then used in the melted salts in exactly the way that has been described for solid lumps of fresh calcium cyanamid. Arrangements have been made with the manufacturers of calcium cyanamid to select high-grade cyanamid for the purposes of this process, and to transfer it directly from the furnace in lumps of desired size to air-tight containers, so that its use in the process will always be at a maximum efficiency.

The quantity of cyanamid immersed may vary according to the size

of the bath and the shape, size, and character of the articles to be case-hardened. A bath may have immersed in it 5 per cent. of its weight of the fresh calcium cyanamid, or a corresponding quantity of the cyanamid coke, for ordinary work.

Upon removal from the bath, the case-hardened articles are quenched in a suitable cooling liquid as in ordinary case-hardening practice.

A careful estimate of the relative cost of running with the sodium cyanide at normal market prices and specially selected calcium cyanamid as described, extending over more than a month's work in a large plant, shows the cost of the bath material to be materially less when using calcium cyanamid than when using sodium cyanide, with the case-hardening done in an equally satisfactory manner, and with much more comfortable and healthful conditions to the workmen.

The question of the scientific basis of the process is being investigated by the inventor, and may form the subject for further communication to the Institute.

DISCUSSION

THE CHAIRMAN (J. E. JOHNSON, JR., New York, N. Y.).—It may be of interest to learn that sodium cyanide is now being made commercially by melting together in an electric furnace calcium cyanamid with common salt, which produces calcium chloride and sodium cyanide. It seems to me entirely possible that you have here about the same results that would be obtained by melting calcium chloride and ordinary cyanide together, only you obtain it a good deal cheaper. There is no doubt about that reaction once it is being carried on on a large industrial scale for the manufacture of cyanide for other uses. I think probably the same thing takes place here, with the possible exception of lowering the melting point due to the presence of the calcium chloride.

J. W. RICHARDS.—The thought of Mr. Landis, of the Cyanamid Co., is that probably sodium cyanamid is produced and goes into the bath. There is no direct evidence that there is cyanide in the bath, but there is possibly some sodium cyanamid, which is an easily decomposing material and probably gives up its carbon to the steel as it decomposes in the bath.

The process has been in practical use for over a year. The salts, however, should be carefully looked after; that is, the calcium chloride of commerce, with its possible impurities, should be carefully selected so as to see that the right quality is obtained. Mr. Shimer has found that using any calcium chloride obtainable will hardly do: it should be a carefully prepared calcium chloride. The American Cyanamid Co. is, I believe, making an electrically calcined calcium chloride that has rather peculiar qualities, such as not readily absorbing moisture

from the air. We do not know yet whether this material is as good a constituent of the bath as the ordinary good-quality calcium chloride of commerce. With a good quality of commercial calcium chloride there is no difficulty about making the right kind of bath. The theory of the operation, when it is worked out, is going to be very interesting, but we do not at present know exactly what it is.

P. W. SHIMER, Easton, Pa. (written discussion*).—When lump calcium cyanamid is immersed in the melt of mixed chlorides of the alkalies and alkaline earths in the manner described in the paper, the amount of sodium cyanide formed in the melt, as shown by numerous analyses, is but a small fraction of that formed by the reaction referred to by Mr. Johnson. The case-hardening effects in my process are, however, at least as good as when the common cyanide baths are used. Moreover, when compared with the cyanide baths, the gases evolved in the new process are harmless, consisting of nitrogen, carbon monoxide (which burns to carbon dioxide on the surface of the melt), and occasionally a little carbon dioxide. These gases are not irritating or poisonous, whereas the vapors arising from cyanide baths are both irritating and poisonous.

The complex chemical reactions that take place in and on the surface of the melt are being investigated by the writer, who hopes, in due time, to communicate the results of his work.

The much lower melting point [about 250° F. (121° C.) lower] of the mixed chlorides, their cheapness, and their freedom from deterioration after long use in the melting pot, are important advantages. The freedom from deterioration of the bath is due to the fact that the melt simply combines with and dissolves case-hardening compounds from the lump cyanamid, leaving all residual lime, carbon, etc., in the lumps, which are lifted out with the basket when the cyanamid is exhausted.

HERBERT HAAS,† San Pedro, Calif. (written discussion‡).—Professor Richards speaks of two essentially different types of case-hardening processes; "that using a dry mixture in which the object to be case-hardened is packed and kept for the necessary time at the necessary temperature, and the "liquid" process, employing a bath of fused salts into which the object is immersed and which by immediate contact case-hardens the surface of the article. The Shimer process belongs to the second class, and was a war emergency invention."

To these two types should be added a third, which hardens by the use of gas; this is the process developed by Adolph W. Machlet, of New Jersey. In this process the objects to be case-hardened are placed in a revolving retort and exposed to a carbon-charged atmosphere under

* Received May 3, 1919.

† Research Engineer, Shipbuilding and Dry Dock Co.

‡ Received Sept. 29, 1919.

pressure. The retort is evenly heated from the outside by a fuel gas and air and the products of combustion have no access to the objects being hardened. The absorption of carbon begins as soon as the objects are sufficiently heated to attract it. As the work is uniformly exposed to a uniform heat in the presence of an abundant supply of volatile carbon, which is admitted as fast as the waste carburizing gas is spent and removed, great speed of operation with uniform results in the carburizing is secured. The value of this process is that the gas will have access to holes, slots, cavities, or any part of the objects not likely to be touched by solid material in the dry case-hardening process, and the greater speed with which articles can be hardened.

A gas-generating plant for this process consists of a petroleum storage tank, a generator, scrubber, and gas producer. The gas is made in the producer, scrubbed, and goes to the generator, where it is charged by carbon vapor derived from refined petroleum, and supplied to the work in the retort as fast as it can be absorbed without forming obstructive carbon deposits. For nearly 10 years, though, the process has been successfully carried out without the use of a special gas-generating plant. It has been found that ordinary city or natural gas can be employed with the same success with considerably less expense. The retort is heated by a fuel gas, for which work either natural or any manufactured fuel gas can be supplied from an extraneous source.

The time of carburizing varies with the material heated and the depth of the case; after the first hour it takes proportionately longer to obtain additional thickness of the case. However, the case after the first two hours is usually thicker than that obtained after several hours of packing in dry material. Heat treatment given to the material after it is carburized also determines final results. A slow cooling (to avoid crystallization of the case), with subsequent reheating at a proper temperature to harden the case, gives better results than if the material were quenched immediately after being removed from the retort. The heat of the retort must also be adapted to the material being treated, as ascertained by trial. The rate of carbon absorption increases with the temperature. High heats tend to crystallize the cases; 1500° F. should not be exceeded before the safe temperature that can be used without detriment to the inner structure of the metal has been ascertained, as indicated by a fractured specimen of the material treated.

This process makes it possible to use for a variety of work, cheaper grades of low-carbon steels, properly case-hardened and heat-treated, instead of the more expensive grades of high-carbon tool steels, in addition to being suitable for the hardening of all work now treated by the "dry" method.

Manufacture and Properties of Light-wall Structural Tubing

BY H. J. FRENCH,* MET. E., WASHINGTON, D. C.

(Chicago Meeting, September, 1919)

WITHIN the past few years, particularly because of the rapid growth of the airplane industry during the war, considerable attention has been paid to the manufacture of light-wall cold-drawn seamless and welded steel tubing for structural purposes. While the manufacture of such tubing is not, in itself, a difficult matter, some difficulties have been encountered in producing the same under specifications requiring definite physical properties. It is the purpose of this paper to describe the methods of manufacture and the properties that may be obtained by such methods for tubing coming within about ½ to 2 in. (12.7 to 50.8 mm.) outside diameter and 10 to 22 Birmingham wire gage (0.134 to 0.028 in. wall thickness). Such tubing is largely used as structural material in the aircraft industry, and it is of the utmost importance that the physical and chemical properties should be known, and also that the various factors entering into the manufacture should be under such control as to allow these properties to be varied and held within definite limits.

In general, tubing may be divided into four classes: namely, *specialty tubing*, such as riveted-sheet or plate, lock jointed, etc.; *brazed tubing; welded tubing;* and *seamless tubing*. Welded tubing may be divided into *hot product*, such as lap welded and butt welded, and *cold product*, such as acetylene and electric welded. Seamless tubing may be hot rolled or cold drawn.

SPECIALTY TUBING

The first two classes present mainly mechanical or structural engineering problems. A simple form of tube is made by riveting together the overlapping edges of rolled sheets or plates formed into a cylinder. Because of the nature of its construction, such a tube is necessarily limited to relatively large sizes. Lock-jointed tubing is produced by passing low-carbon strip through a forming machine having suitable rolls and dies, its manufacture is almost entirely a question of design of the forming machine and its satisfactory operation. The chief application of such tubing is in the manufacture of bedsteads.

*Chief of Heat-treatment Section, Division of Metallurgy, U. S. Bureau of Standards.

Brazed Tubing

Production of brazed tubing presents no special difficulties, though the application of metallurgical principles is involved in the application of a correct brazing mixture and proper heating. A simple method of manufacture consists in forming the tube from strip in a machine having a series of rolls with properly designed holes of successively decreasing radii and then passing the formed strip at a uniform rate under a funnel-shaped device, which feeds the brazing mixture directly onto the line where the two edges of the strip meet. The tube is later passed endwise through a heated muffle by means of grooved racks traveling on an endless chain, which passes through the heated chamber and, on the return, is passed through a trough of water for cooling.

Seamless Tubing

The manufacture of seamless steel tubing, the most important type for structural purposes, is of comparatively recent origin, beginning little more than 25 yr. ago. Its development has been accelerated by the bicycle, the automobile, and the airplane, the latter creating a demand for a most refined and closely controlled product.

The conversion of a solid piece of steel into a seamless tube may be accomplished by boring a hole in the center of a round steel bar, by forming sheet or plate between a series of suitable dies, and by piercing by the Mannesmann process or one of its modifications.

The first and oldest method consisted in merely boring a hole through the center of a long round bar; besides being a laborious operation and very difficult to perform, this was expensive. Various modifications to simplify this method of production may be used. A hole may be bored through the center of a short bar, which may then be drawn out cold; or the hollow bar may be placed in a hydraulic press and the hole enlarged by forcing into it successively larger rams after which the tube may be rolled out over a mandrel in a manner similar to that by which lap-welded pipe is rolled. These modifications are easier to carry out than the original method and much less expensive. Other modifications may also be used depending on the character of the product desired.

The second method is used for the manufacture of gas containers and consists in pressing a heated flat circular plate between a series of dies. In this process the depression successively becomes deeper until finally the sides of the cup formed are parallel. For light-gage sheet, this cupping may be performed cold in three or four stages and the tube thus formed, with one closed end, may be finally drawn out in the ordinary manner, which will be described later.

Mannesmann Process of Making Seamless Tubing.—The most important method of producing seamless tubes is the Mannesmann process or modifications of the principles involved. The underlying principles of this process were first applied by the Mannesmann Brothers, German tool-steel manufacturers, and today practically all light-wall, cold-drawn, seamless, steel tubes used in the aircraft industry in America are made from hot-rolled billets pierced in this way.

Hot-rolled rounds of suitable composition, usually from 3 to 6 in. (7.5 to 15 cm.) in diameter, commonly called billets, are hot sawed into proper lengths depending on the length of tube desired. These lengths are then heated in ordinary heating furnaces and fed to a pair of conical rolls set at an angle to each other, as shown graphically in Fig. 1. The bar, which is at a bright red heat, is rapidly rotated as it enters the small diameter end of these rolls, which exert a pressure on two constantly changing points, but which are always on opposite sides of the center, or

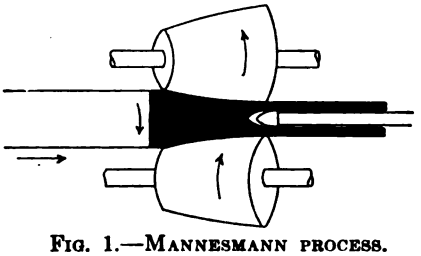


FIG. 1.—MANNESMANN PROCESS.

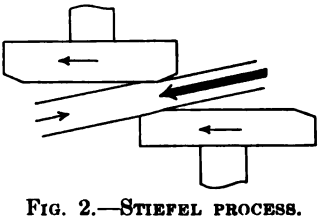


FIG. 2.—STIEFEL PROCESS.

axis, of the bar. During the first few inches of longitudinal travel imparted to the bar, a flaw, or point of weakness, at or near the center is formed. The action of this part of the rolls may be likened to the effect of pressure exerted by flat dies in a hammer when applied to a round bar heated to normal working temperatures. Under such conditions, there is a tendency for the metal to spread outward from the center instead of being compressed by the blow. As the bar is pulled forward between the rolls, it reaches the section of larger diameter, where the linear speed is greater than at the small end of the rolls. There is, therefore, a tendency to twist the outer surface of the bar, which is also pulled forward. This intensifies the weakening effect at the center of the bar, and as the bar is lengthened more rapidly than it decreases in diameter, metal must be drawn from somewhere—which will naturally be the weakest point or the center of bar. As the bar is leaving the rolls, it is forced over a piercing head at the end of a strong heavy rod which completes the formation of the tube. The tube is removed from this rod, and a new head is placed on the same preparatory to piercing the next bar. Because of the shape of the rolls, this equipment is often called a barrel-type mill.

Stiefel Process.—The disk-type mill, or Stiefel process, graphically represented in Fig. 2, is also used in forming tubes by the application of external pressure.

The pierced billet, or tube formed, is rough and has heavy irregular walls. Further reduction is therefore necessary, so the tube is again heated and fed to a pair of rolls which are grooved as for rolling round bars. In back of each opening is a mandrel, which has a loose nose and is carried by a strong cross-bar attached to the housings at the back of the stand. As the tube is passed through the rolls and forced over the mandrel, it is reduced in outside diameter and wall thickness, the latter being determined by the size of the mandrel. This first stand is usually referred to as the breaking-down rolls. The tubes are then usually reheated and passed to the finishing rolls; if further reduction is desired, they may be passed directly to a reducing mill, consisting of a series of rolls alternately rotating about horizontal and vertical axes. The size of hole from the first to the last set of rolls decreases so that the tube is decreased in outside diameter.

The hot-rolled pierced billets are reduced to varying sizes and wall thicknesses down to about 1¾ in. (4.5 cm.) outside diameter and ⅛ in. wall. They are next passed to the cold-drawing mill, where they are pickled, washed and dried, drawn, annealed, and again pickled, the cycle of operations being repeated as often as necessary to produce the desired size and gage in the cold-drawn tube.

For pickling, a dilute hot acid is used. In place of the more generally used sulfuric acid, some plants use solution of niter cake, which is a byproduct from the manufacture of nitric acid. When hot, this solution does not give off the obnoxious fumes obtained from the use of sulfuric acid and the relatively weak concentration of the acid radical decreases the danger of over-pickling. In either case, some of the scale is dissolved and the remainder loosened so that it readily comes off in washing. After removal from the pickling bath, the tubes are washed, dipped into lime water, and dried. Before being cold-drawn, they are pointed at one end by means of a drop hammer or other device.

The draw-bench consists, primarily, of a long heavy frame on which is mounted, by suitable means, an endless chain, which extends from one end to about the center and which slowly travels toward that end. The tube is threaded over a mandrel or bar and the pointed end is passed through the die, which is attached to the frame near the center of the bench. The pliers of the draw box are attached to the pointed end of the tube, the box is attached by a hook to the slowly moving endless chain, and the tube is drawn out, as shown in Fig. 3. Grease or tallow is used as a lubricant.

Cold-drawn Seamless Tubing.—In cold drawing a tube, one of three methods may be used: (1) Drawing through a die without the use of

mandrel or rod, called "sinking;" (2) drawing over a mandrel and through a die, referred to as the "back-bench method;" (3) drawing on a rod through the die. The first method obviously reduces only the diameter of the tube. The second and third methods accomplish the reduction of the outside diameter and the wall thickness.

In the back-bench method, the tube is pulled over a mandrel at the end of a long bar attached to that end of the draw bench from which the tube is pulled. As the mandrel is fixed, both the inner and outer surfaces of the tube become friction surfaces in the drawing, and the tube takes the entire "pull," which is limited to that force which would break the thin wall of the tube.

In the third method, the tube is threaded on a rod that has been cold-rolled, drawn, or machined to a uniform diameter, and the two are drawn together through the die. In this case the tube has but one friction surface, and the "pull" is taken by the bar of very much greater cross-sectional area than the thin-wall tube, which sustains the force applied by the second method. As would be expected, greater reduction

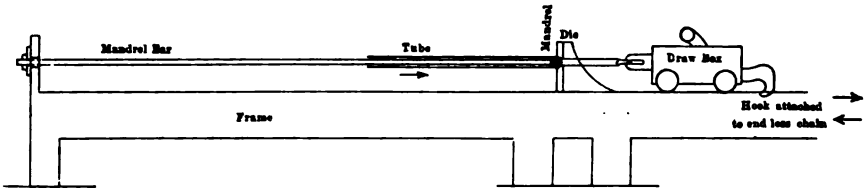


FIG. 3.—DRAW BENCH.

per pass can be used when drawing on a rod, and it is possible to reduce a tube a given amount in outside diameter and wall in fewer passes than by the use of the back-bench method.

After drawing, the tube is tightly attached to the rod, so special means must be taken to remove it. This may be done by passing the rod and attached tube a large number of times through a pair of rolls having a groove of larger radius than the outside diameter of the tube. With each pass through the rolls the tube is rotated slightly so that the line of contact is changed. The pressure applied lightly reduces the wall thickness of the tube, which is accompanied by expansion, so that it is possible to remove the parallel mandrel with ease. Another method of releasing is to pass the tube through a "reeling" machine, which consists primarily of a pair of conical rolls revolving in the same direction, the axes of the rolls being inclined to the horizontal in opposite directions. The details of this machine will not be given, its action being to slightly reduce the wall thickness, which is accompanied by expansion in the diameter. After one pass the mandrel may readily be removed.

Testing and Inspecting Seamless Tubing.—After the final pass, which brings the tubes to the desired size and gage, they are annealed at a comparatively low temperature, depending on the analysis, amount of cold work, and desired properties. They are then straightened and the ends cut off; if special lengths are desired, they are cut to these lengths and inspected. The inspection made varies considerably, depending on the specifications, which, in turn, are based on requirements for the particular purposes for which the tubes are desired. For the type of tubing under consideration, the inspection may be considered to consist of a thorough visual examination of the tubes, including a check on the size and gage and the determination of the chemical and physical properties.

For physical test, a number of tubes from a batch are chosen and a sample about 14 in. (35 cm.) long is cut from each tube. Usually one sample tube is chosen for each 400 ft. (122 m.) of tubing presented, though this number is greatly increased when a question regarding the uniformity of the lot arises. Marks 1 in. (2.5 cm.) apart are scribed over practically the entire length of the sample along two lines approximately 60° around the circumference of the tube, which is placed in the testing machine after the insertion of slightly tapered solid plugs in each end. The pulling head of the testing machine is brought to such a position that the jaws take hold of the tube with the contained plugs at a distance of about 8 in. from each other and the load is applied at a uniform and slow rate, not exceeding 0.50 in. per min. The yield point is obtained by calipers scribing a line of about 5 in. radius and noting the load causing a permanent set. This method requires the release of the load and its reapplication; but when experienced the operator can usually make this determination without releasing the load more than once. Having determined the yield point, the rate of application of load is increased to about 0.80 to 1.0 in. per min. to break the tube. The elongation is measured in the usual manner, except that the maximum readings of the two original 2-in. and several possible 8-in. gage lengths, including the break, are the values taken.

While the elongation in the tensile test is measured, two other tests are largely used as a measure of the ductility. The compression test consists in the application of pressure parallel to the axis of the tube on a sample whose length is 1½ times the outside diameter. Specifications usually require that the tube, in such a test, shall not show cracks when the diameter has been increased to a given per cent. or when one complete fold has formed. However, the test is often carried beyond this specified expansion to the point of failure, which is measured in per cent. increase of the original diameter. The second test for ductility is the flattening test, in which pressure is applied to the tubes perpendicularly to the axis of the tube until the walls are at no greater distance from each

other than twice the thickness of the wall of the tube; under such conditions the tube must show no cracks. Fig. 4 shows the character of these tests. Obviously the compression test is more a measure of the ductility of the tube at right angles to the direction of cold work while the flattening test is a measure of the ductility with the grain of the metal.

As in cold-drawing, ten to fifteen passes are not uncommon in producing tubes of thin wall and small diameter, rigid inspection is necessary to guard against inclusion of defective material in any given lot. The importance of this will possibly become clearer after a somewhat de-

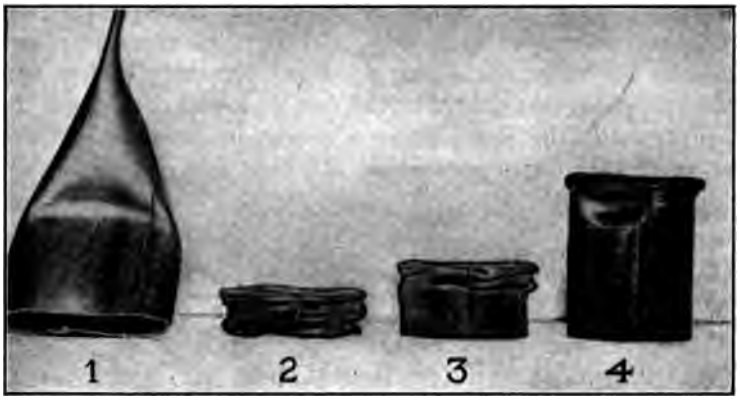


FIG. 4.—RESULTS OF FLATTENING AND COMPRESSION TESTS. 1, FLATTENING TEST MADE ON A WELDED STEEL TUBE; 2, 3 and 4, DIFFERENT STAGES IN COMPRESSION TEST MADE ON A SECTION OF WELDED STEEL TUBE; 2, COMPLETE COMPRESSION; 3, THREE COMPLETE FOLDS; 4, COMPRESSION FOR GIVEN PERCENTAGE EXPANSION IN DIAMETER. NOTE FAILURE IN THE WELD OF ALL TESTS SHOWN.

tailed consideration of the effect of various factors entering into the manufacture of the different grades of structural tubing in common use in the aircraft industry.

SEAMLESS TUBING USED IN AIRCRAFT

This type product is largely used in stressed parts in airplane construction and in all probability will be more extensively employed in the future. The three main variables by which the properties of the tube of given size may be controlled are: Chemical composition, cold work, and thermal treatment.

Three type compositions are generally produced and most widely used: (1) Low-carbon steel, usually 0.10 to 0.20 per cent. carbon or 0.15 to 0.25 per cent. carbon open-hearth; (2) medium low-carbon steel, 0.30 to 0.40 per cent. carbon open-hearth; (3) medium low-carbon, 3.5 per cent. nickel steel, 0.25 to 0.35 per cent. carbon, 3 to 3.50 per cent.

nickel open-hearth. As the second and third type compositions are generally used in the heat-treated condition, they will be referred to later. The increase in properties obtained by the heat treatment of low-carbon steel is not commensurate with the cost of the operation, particularly when two such controlling factors as cold work and simple annealing can be used for regulation of these properties. This type composition is, therefore, generally used in the unheat-treated condition, or as drawn.

Chemical Composition.—Regardless of the type composition chosen, the first requisite for the production of tubing of uniform properties is uniformity in chemical composition. In Table 1, the results recorded in each series are taken from one lot of tubes, each receiving the same cold reduction and annealing, as noted. Although the plan is not generally practised by the mills, the writer is a firm believer in the necessity for heat identification, principally in the case of low-carbon tubes, to more nearly insure, with the large number of operations involved in the production of such tubes, the prevention of mixed analyses as is illustrated in series 2.

TABLE 1.—*Physical Properties of Low-carbon Seamless Tubes*

Analysis, Per Cent.				Yield Point, Lb. per Sq. In.	Tensile Strength, Lb. per Sq. In.	Per Cent. Elongation in		Compression at		Flattening Test
Carbon	Manganese	Phosphorus	Sulfur			2-In.	8-In.	25 Per Cent.	Failure	
Series 1.—1¼ in. by 12 B.W.G. Drawn from 2.118 by 0.115 wall, annealed at 1020° to 1060° F. for 45 min.										
0.19	0.45	0.028	0.024	47,910	61,060	35.5	17.0	O.K.	Skin cracks at complete compression	O.K.
0.08	0.34	0.028	0.027	41,910	54,450	51.5	23.5	O.K.	.....	O.K.
Series 2.—1 in. by 17 B.W.G. Cold reduction and final annealing not given but same for both tubes										
0.15	.....	.....	.....	45,100	57,400	.....	.....	O.K.		
0.33	.....	.....	.....	94,500	108,300	.....	.....	Badly cracked		

Cold Work and Thermal Treatment.—For tubing of uniform and predetermined chemical properties, there are two variables, the cold work and the thermal treatment, that may be varied so as to regulate the final properties. Cold work increases the yield point and tensile strength but decreases the ductility, as measured by the elongation, reduction in area, compression, and flattening tests. Reduction of the wall of the tube, whether by the back-bench method or by the method of parallel mandrels, is more effective in increasing the yield and tensile strength and inducing brittleness than the reduction of the diameter by "sinking."

Annealing lowers the yield point and tensile strength but greatly

increases the ductility, as measured by the various factors mentioned in the preceding paragraph. It is, therefore, evident that with these two factors having opposite effects on the physical properties, it is possible to vary these properties within certain limits by regulation or control of the variables. Tables 2 and 3 show the quantitative effect of cold work and annealing on low-carbon steel tubing.

TABLE 2.—*Effect of Varying Cold Reduction in Outside Diameter on Seamless Steel Tubing*

0.10 to 0.20 per cent. carbon steel

Tube Mark	Treatment	Measured, Inch		Yield Point, Lb. per Sq. In.	Tensile Strength, Lb. per Sq. In.	Per Cent. Elongation in		Compression at		Flattening Test
		Outside Diameter	Wall			2 In.	8 In.	25 Per Cent.	Failure	
A1	As annealed, before last pass	1.740	0.069	27,875	47,475	55.0	32.0	O.K.	No cracks at complete compression.	O.K.
2	1 reduced 1/16 in. and 1 gage, as drawn	1.597	0.054	43,180	51,970	45.5	20.5	O.K.		O.K.
3	As for sample 2, then annealed at 980° F.	1.597	0.055	36,790	51,250	46.0	24.5	O.K.		O.K.
4	1 reduced 3/16 in. and 1 gage, as drawn	1.545	0.055	42,450	55,910	35.0	13.5	O.K.		O.K.
5	As for sample 4, then annealed at 980° F.	1.546	0.054	41,960	56,750	40.5	20.0	O.K.		O.K.
B1	As annealed, before last pass	1.734	0.069	28,110	47,120	57.0	32.0	O.K.	No cracks at complete compression.	O.K.
2	1 reduced 1/16 in. and 1 gage, as drawn	1.597	0.054	42,760	52,660	44.5	21.0	O.K.		O.K.
3	As for sample 2, then annealed at 980° F.	1.597	0.054	37,450	51,970	47.5	25.5	O.K.		O.K.
4	1 reduced 3/16 in. and 1 gage, as drawn	1.546	0.055	48,450	58,160	41.0	13.0	O.K.		O.K.
5	As for sample 4, then annealed at 980° F.	1.546	0.055	41,660	55,980	42.0	19.0	O.K.		O.K.
	Average of two tubes									
1	As annealed before last pass			27,992	47,297	56.0	32.0			
2	1 reduced 1/16 in. and 1 gage, as drawn			42,970	52,315	45.0	20.75			
3	As for sample 2, then annealed at 980° F.			37,120	51,610	46.75	25.0			
4	1 reduced 3/16 in. and 1 gage, as drawn			45,450	57,035	38.0	13.75			
5	As for sample 4, then annealed at 980° F.			41,810	56,365	41.25	19.5			

NOTE.—Annealing at 980° F. of the tubes listed was performed at one time in one furnace to avoid any undue discrepancies.

It is at once evident that for uniformity in physical properties, tubes of varying degrees of cold work on the last pass should not be included in the same lot. A different annealing temperature would be required, as an example, for a tube finished by sinking 1/16 in. (1.5 mm.) to a given

TABLE 3.—*Effect of Varying Wall Reduction on Last Pass in Producing Low-carbon Steel Tubing*

(0.10 to 0.20 per cent. carbon steel. 0.654 in. outside diameter × 0.036 in. wall)

Tube Mark	Treatment	Yield Point, Lb. per Sq. In.	Tensile Strength, Lb. per Sq. In.	Per Cent. Elongation in 2 In.	Per Cent. Elongation in 8 In.
A1	As annealed, 0.654 in. outside diameter × 0.036 in. wall.....	23,605	49,350	40.0	26.0
2	1 reduced 1/32 in. in diameter and 0.008 in. in wall thickness.....	66,930	70,000	6.0	2.0
3	As for sample 2, then annealed at 1000° F.....	55,260	66,540	24.0	14.5
B1	As annealed, 0.654 in. outside diameter × 0.036 in. wall.....	24,320	49,640	46.5	29.0
2	1 reduced about 1/32 in. in diameter and 0.005 in. in wall thickness.....	55,240	56,900	7.5	2.5
3	As for sample 2, then annealed at 1000° F.....	40,420	51,650	24.5	12.5

TABLE 4.—*Effect of Low-temperature Annealing on Properties of Low-carbon Cold-drawn Seamless Steel Tubes*

Tube	Size and Gage (B.W.G.)	Approximate Reduction from, Inch	Approximate Final Annealing Temperature, Degrees F.	Yield Point, Lb. per Sq. In.	Tensile Strength, Lb. per Sq. In.	Per Cent., Elongation in 2 In.	Per Cent., Elongation in 8 In.	Remarks
3	1 1/2 × 10	1 5/16 × 9	As drawn	66,050	69,800	17.5	6.5	Results on 0.10 to 0.20 per cent. carbon steel.
3			900	49,900	69,000	24.0	11.0	
4	1 1/2 × 10	1 5/16 × 9	As drawn	61,500	68,200	14.5	7.5	
4			900	62,950	71,400	28.0	12.0	
1	1 3/4 × 10	1 13/16 × 9	As drawn	60,500	65,700	28.0	10.0	
1			900	58,750	69,500	31.5	14.5	
2	1 3/4 × 10	1 13/16 × 9	As drawn	65,700	69,500	29.0	8.5	
2			900	49,950	60,450	36.5	17.5	
1A	3/8 × 10	1 5/16 × 9	As drawn	48,200	58,700	13.0	9.0	
1A			900	52,400	62,600	27.5	15.5	
2A	3/8 × 10	1 5/16 × 9	As drawn	53,800	68,800	30.5	13.5	
2A			900	66,700	77,600	23.0	13.0	
21	1 3/4 × 14	1.8 × 0.100	As drawn	58,600	65,750	31.5	8.5	
21			900–950	57,600	68,000	26.5	12.0	
22	1 3/4 × 14	1.8 × 0.100	As drawn	63,200	69,300	21.0	6.5	
22			900–950	64,700	73,700	26.0	12.0	
33	0.790 × 17	0.840 × 16	As drawn	47,650	71,250	15.5	5.0	
			900–950	63,750	76,500	20.0	12.0	
5	1.45 × 20	1.50 × 19	As drawn	81,000	81,750	16.0	5.0	
5			900–950	70,500	80,000	22.5	13.0	
6	1.45 × 20	1.50 × 19	As drawn	77,600	84,700	18.0	5.5	
6			900–950	67,500	77,650	17.5	11.0	
7	0.660 × 20	0.720 × 19	As drawn	59,000	75,500	9.5	3.0	
7			900–950	64,250	75,000	18.0	10.0	
8	0.660 × 20	0.720 × 19	As drawn	51,500	54,850	20.5	9.0	
8			900–950	50,300	60,800	21.5	13.0	

size and a tube finished with a reduction of ⅛ in. (3.1 mm.) and 0.010 in. (0.25 mm.) to the same size in order to bring out similar properties. It is noted that the greater the cold reduction applied, the greater are the increase in strength factors and decrease in ductility, which is as expected.

TABLE 5.—*Effect of Varying Temperature of Annealing on Properties of Low-carbon Cold-drawn Seamless Steel Tubes*

Tube Mark	Condition in which Tested	Yield Point, Lb. per Sq. In.	Tensile Strength, Lb. per Sq. In.	Per Cent. Elongation in 2 In.	Per Cent. Elongation in 4 In.
0.10 to 0.20 per cent. carbon steel, ¾-in. by 18 B.W.G. drawn from 0.790-in. by 0.055-in. wall after being thoroughly annealed.					
A1	As drawn	74,340	78,270	12.5	7.0
2	As drawn and then annealed at 700° F.	71,210	80,350	17.0	12.5
3	As drawn and then annealed at 800° F.	69,510	79,620	19.0	13.2
4	As drawn and then annealed at 900° F.	63,090	75,870	19.0	14.0
5	As drawn and then annealed at 1000° F.	59,050	72,280	25.5	18.0
B1	As drawn	75,910	78,430	12.5	7.5
2	As drawn and then annealed at 700° F.	70,700	78,420	16.5	11.7
3	As drawn and then annealed at 800° F.	67,930	79,300	18.5	13.5
4	As drawn and then annealed at 900° F.	63,470	76,380	20.0	15.0
5	As drawn and then annealed at 1000° F.	58,270	71,020	24.0	16.2
0.10 to 0.20 per cent. carbon steel, ¾-in. by 22 B.W.G. drawn from 0.755-in. by 0.031-in. wall.					
C1	As drawn	34,850	48,000	45.0	34.5
2	As drawn and then annealed at 700° F.	41,220	51,080	41.5	30.7
3	As drawn and then annealed at 800° F.	34,640	46,980	42.5	33.7
4	As drawn and then annealed at 900° F.	33,400	46,400	43.0	34.5
5	As drawn and then annealed at 1000° F.	32,350	46,720	45.0	34.0

NOTE.—Annealing consisted in heating to the designated temperature in 15 min., holding at this temperature for 15 min. and cooling in air.

When comparatively low temperature is used for annealing low-carbon steel tubes, the yield point and tensile strength are only slightly affected, though the ductility, as measured by elongation and compression tests, is immediately increased. Often the increased yield point and tensile strength due to cold work are raised slightly; in some cases the increase in ductility has been accompanied by practically no change in the yield point and tensile strength while at other times they have been lowered but slightly, certainly not in proportion to the increase in ductility effected.

Table 4 shows the results of a few experiments along these lines on tubes of varying sizes and gages. As the temperature is further increased, the yield point and tensile values fall with corresponding increase in ductility, the degree depending on the temperature, etc.; Table 5 illustrates this fact.

TABLE 6.—*Effect of Furnace Temperature Variation in Final Annealing on Properties of Low-carbon Seamless Steel Tubing*

Size and Gage (B.W.G.)	Temperature as Given by Pyrometer, Degrees F. (Final Anneal)	Part of Furnace from which Chosen by Observer	Yield Point, Lb. per Sq. In.	Tensile Strength, Lb. per Sq. In.	Per Cent., Elongation in	
					2 In.	8 In.
0.276×0.039 wall..	1140–75	Hottest	37,700	57,500	36.5	25.5
		Coolest	55,700	67,000	22.0	11.0
		Hottest	35,900	51,500	37.0	21.5
1⅜×17.....	1000–1040	Coolest	53,300	65,500	30.5	15.0
		Hottest	59,700	68,500	31.0	15.5
		At random	51,200	62,400	31.0	15.5
		At random	60,000	70,300	32.5	16.0
0.394×0.079 wall..	1140–75	Hottest	35,000	53,800	43.0	24.5
		Hottest	39,100	57,600	34.5	17.5
		Coolest	50,600	61,300	26.0	13.0
		Coolest	55,600	65,000	23.0	11.0

Furnaces that can be closely controlled within narrow limits of temperature are not only desirable but necessary in annealing tubing if the desired uniformity in physical properties is to be obtained. While the preceding tables shows the effect of temperature variations in final annealing, Table 6 will further confirm this statement as there are included results on tubes picked from various parts of a furnace which, according to the observer, lacked the required uniformity.

Objections to Cold-drawn Tubing.—With the foregoing in mind, it will be well to consider what may happen in actual mill operations in the production of thin-wall tubing. The writer has several times heard the statement made that it is impossible to produce such tubing having very uniform properties because of the marked effect of small variations in mechanical reduction bound to occur due to wear in dies and mandrels and because small variations in the wall of a tube having a very small cross-sectional area represent a high percentage change in this area, which may greatly affect the results obtained in actual pounds yield point and tensile strength and, therefore, also the values in pounds per square inch. With careful measurements in testing and properly designed specifications covering tensile properties, the writer does not believe the second objection holds good for 20 gage and heavier walls. Certainly a real problem is presented for finer sizes, such as 24 B.W.G. If specifications

are properly drawn up it is believed satisfactory tubing can be produced even in these very thin gages.

Regarding the first objection let it be assumed that ¾ in. (19 mm.) by 20 B.W.G. tubing is being produced from ⅞ in. (22 mm.) by 18 B.W.G. in two passes, as follows: ⅞ in. × 18 B.W.G., $^{13}/_{16}$ in. × 19 B.W.G., ¾ in. × 20 B.W.G.

The first tube *A* drawn with a new mandrel 0.002 in. over-size (0.7305 in.) and a new die 0.002 in. under-size (0.8105 in.) receives a reduction of 0.0625 in. in its outside diameter and 0.009 in. in the thickness of the wall. As the drawing proceeds the die wears and becomes 0.003 over-size (0.8055 in.) and the mandrel becomes 0.002 in. under-size (0.7265 in.). The last tube of the lot drawn, *B*, has then only received a reduction in diameter of 0.0595 in. and in wall thickness of 0.0045 in. The tubes are uniformly annealed and then redrawn from approximately $^{13}/_{16}$ in. by 19 B.W.G. to the final size ¾ in. by 20 B.W.G. This time tube *B* is the first to be drawn and *A* is the last. The latter receives a reduction of only 0.0605 in. in diameter and 0.005 in. in wall while the former, *B*, receives the heavier reduction of 0.0655 in. in diameter and 0.0095 in. The physical properties, as expected, will vary considerably after each of these passes. If in addition the intermediate annealing lacked uniformity and tube *A* was heated to a higher temperature than tube *B* even wider variation would be expected and obtained. The tubes are then finally annealed at a low temperature, which tends to equalize to some extent the differences created by the cold drawing, but the desired uniformity is still lacking. With this method of finishing tubes, considerable variations in physical properties may be obtained.

If in the case cited tube *B* had been drawn in a new die, several thousandths of an inch below 0.750 in. and over a new mandrel several thousandths of an inch over-size and tube *A* had been drawn after wear in both die and mandrel, the variations in reduction and, therefore, variations in physical properties would become more marked.

Such conditions naturally represent extremes, but may occur. The results of an interesting experiment to determine the effect of such varying wall reductions on the physical properties of three tubes are given in Table 7. The measurements were carefully taken by micrometers and the values obtained in the tensile tests were all taken by an experienced operator. The tubes chosen for this experiment were finished in the usual manner by drawing over a mandrel in the last pass and then annealing at 1000° F. Tests were made on each tube which was then annealed at a fairly uniform and higher temperature approximating the usual interpass anneal for further relief of previous drawing strains. The tubes were then given various reductions as shown in the table.

Disadvantages of Finishing by Cold Drawing.—Obviously the method of finishing outlined is one that requires that minimum wear be allowed

TABLE 7.—*Effect of Varying Wall Reduction (Such as is Met in Ordinary Drawing Practice) on Medium Size 0.10 to 0.20 Per Cent. Carbon Seamless Steel Tubes*

(Three tubes A, B, and C, 1½ × 16 B.W.G., annealed at 1000° F.. 30 min., were drawn from 1¾ × 16 B.W.G. after being annealed at 1380–1400° F. for 10 min.)

Tube	Sample		Inside Diameter, Inch	Wall Thickness, Inch	Reduction in		Yield Point, Lb. per Sq. In.	Tensile Strength, Lb. per Sq. In.	Per Cent. Elongation in		Compression		Flattening Test
					Outside Diameter, Inch	Wall Thickness, Inch			2 In.	8 In.	25 Per Cent.	Complete	
A	1	1.499	1.377	0.061	..	.....	53,380	65,135	27.5	15.0	O.K.	O.K.	Bad cracks
B	1	1.499	1.373	0.063	..	.....	56,280	68,240	22.5	12.5	O.K.	O.K.	O.K.
C	1	1.498	1.377	0.061	..	.....	57,510	69,260	31.0	17.0	O.K.	O.K.	O.K.
					After these tubes were annealed at 1380–1400° F. for 10 min.								
A	2	1.499	1.373	0.063	..	.....	21,525	46,220	58.0	34.0	O.K.	O.K.	O.K.
B	2	1.500	1.374	0.063	..	.....	25,915	49,050	56.5	34.0	O.K.	O.K.	O.K.
C	2	1.499	1.377	0.061	..	.....	24,310	50,730	53.5	31.5	O.K.	O.K.	O.K.
					These tubes were then drawn using different wall reductions as given in the following table								
A	3	1.375	1.255	0.060	1/16	0.003	43,665	57,610	39.0	21.0	O.K.	O.K.	O.K.
A	4	1.375	1.260	0.058	1/16	0.005	47,270	60,640	33.0	17.0	O.K.	O.K.	O.K.
A	5	1.375	1.266	0.055	1/16	0.008	50,200	62,890	24.0	15.0	O.K.	O.K.	O.K.
B	3	1.375	1.255	0.060	1/16	0.003	48,345	61,140	36.0	19.5	O.K.	O.K.	O.K.
B	4	1.375	1.259	0.058	1/16	0.005	53,250	65,010	20.0	13.0	O.K.	O.K.	O.K. Fin.
B	5	1.375	1.265	0.055	1/16	0.008	57,820	67,960	14.0	8.0	O.K.	O.K.	O.K. Fin.
C	3	1.376	1.256	0.060	1/16	0.001	44,535	59,530	35.0	18.5	O.K.	O.K.	O.K.
C	4	1.375	1.260	0.058	1/16	0.003	51,390	65,000	26.0	14.0	O.K.	O.K.	O.K.
C	5	1.375	1.266	0.055	1/16	0.006	55,290	67,890	12.0	9.5	O.K.	O.K.	O.K.

Above annealing 1000° F. for 30 min. was performed in one furnace charge. The nine samples above were then drawn to 1¼ × 0.050 and annealed 33 min. 1005° F

A	3ª	1.250	1.150	0.050	3/8	0.010	59,760	72,220	26.0	13.5	O.K.	O.K.
A	4ª	1.250	1.150	0.050	3/8	0.008	52,920	65,750	25.0	13.0	O.K.	O.K.
A	5ª	1.251	1.151	0.050	3/8	0.005	48,305	61,395	31.0	16.5	O.K.	O.K.
B	3ª	1.251	1.151	0.050	3/8	0.010	58,790	71,085	24.5	11.5	O.K.	O.K.
B	4ª	1.251	1.151	0.050	3/8	0.008	56,360	69,125	24.0	12.0	O.K.	O.K.ª
B	5ª	1.251	1.151	0.050	3/8	0.005	54,875	68,065	14.0	8.5	O.K.	O.K.ª Def. tube
C	3ª	1.250	1.149	0.050	3/8	0.010	63,555	76,110	23.5	12.0	O.K.	O.K.
C	4ª	1.251	1.151	0.050	3/8	0.008	58,160	70,135	24.0	13.0	O.K.	O.K.ª
C	5ª	1.250	1.150	0.050	3/8	0.005	55,410	69,520	28.5	15.0	O.K.	O.K.ª

ª Repulled for elongation, file mark too deep and broke in this mark originally.

on dies and mandrels used on the last two passes if results of satisfactory uniformity are to be obtained. With a maximum variation of about 15,000 lb. per sq. in. (1054 kg. per sq. cm.), with a tensile strength from 55,000 to 70,000 lb. per sq. in. (3866 to 4921 kg. per sq. cm.), satisfactory ductility can be obtained with 0.10 to 0.20 per cent. carbon tubing. These limits are sufficiently wide for all gages under consideration, though for the heavier walls from 10 to 14 B.W.G. the tensile values may be held within closer limits. However, there are certain disadvantages to this method of finishing by mandrel drawing or sinking followed by a final partial anneal, which are worth consideration.

The tubes are oxidized, sometimes quite heavily, depending on the temperature of annealing and time at heat, etc., and in addition are often out of round in their final form. This latter is particularly noticeable in comparatively large-diameter tubes (1½ to 2 in.) and thin walls such as 18 or 20 B.W.G. Table 8 shows that, in some cases, this variation is beyond reasonable limits and in many cases is certainly more than is desirable. Such tubing is often required to fit into a reamed hole in a casting or forging and while the average outside diameter may be within the prescribed limits in size, the "spring" in the tube makes it impossible to fit the same into the closely reamed hole.

Suggestions for Improving Drawing Methods.—While with careful control in the mill, tubing of satisfactory uniformity in physical properties can be produced, the question naturally arises as to how the procedure may be bettered. The writer believes that the analysis should be so chosen that the cold work is only

TABLE 8.—*Actual Variations in Size of Cold-drawn Seamless Steel Tubing Partly Annealed After Last Cold Pass*

Specified Diameter In.	Specified Wall Thickness, In.	Measured Outside Diameter, Inch				Measured Wall Thickness, Inch				Number of Tubes Measured
		Maximum	Minimum	Average	Maximum Spring	Maximum	Minimum	Average	Maximum Variation	
1.25	0.134	1.255	1.247	1.251	0.008	0.144	0.125	0.133	0.019	6
1.00	0.083	1.004	0.990	0.998	0.014	0.088	0.078	0.083	0.010	6
1.12	0.095	1.131	1.125	1.127	0.006	0.098	0.093	0.096	0.005	6
1.00	0.062	1.008	1.001	1.005	0.007	0.068	0.062	0.065	0.006	6
0.687	0.062	0.693	0.680	0.887	0.013	0.072	0.060	0.066	0.012	6
0.625	0.035	0.635	0.624	0.629	0.011	0.043	0.033	0.039	0.010	6
0.625	0.035	0.632	0.624	0.627	0.008	0.040	0.034	0.038	0.006	6
1.125	0.049	1.127	1.123	1.124	0.004	0.052	0.046	0.049	0.006	6
1.00	0.035	1.025	0.994	1.001	0.031	0.039	0.032	0.036	0.007	6
1.50	0.094	1.508	1.490	1.504	0.018	0.101	0.085	0.096	0.016	6
2.00	0.083	2.001	1.995	1.999	0.006	0.090	0.080	0.085	0.010	6
2.187	0.095	2.202	2.187	2.192	0.015	0.103	0.091	0.096	0.012	6
2.00	0.049	2.005	1.195	2.001	0.010	0.055	0.045	0.050	0.010	6
1.75	0.035	1.757	1.747	1.752	0.010	0.039	0.032	0.035	0.007	6
1.75	0.065	1.760	1.740	1.751	0.020	0.070	0.055	0.067	0.015	6
2.00	0.049	2.012	1.996	2.004	0.016	0.055	0.044	0.051	0.011	6
2.219	0.095	2.234	2.215	2.226	0.019	0.100	0.087	0.093	0.013	6

NOTE.—Readings taken on each tube approximately 60° around circumference. Maximum and minimum values obtained on tube of widest variation recorded. Average is not mean of maximum and minimum, but actual average of three readings taken on all six tubes. Spring is difference between maximum and minimum values.

depended on for close sizing instead of being used to very much increase what would normally be the tensile values in the hot-rolled or annealed state. On next to the last pass, the tubes may receive approximately the reductions now being used, checking closely the wear on the dies and mandrels, but should be brought in this pass to a size and gage close enough to that desired so that after a full anneal the final pass will be more in the nature of sizing with only a slight increase in the tensile strength and yield point due to cold work. This may then be followed by a low-temperature anneal that will not oxidize the surface to any greater extent than to color the same, not lowering the strength but increasing the ductility.

Having a batch of tubes of uniform size and gage just prior to the last pass, close to the desired size and having removed the stresses imparted in previous cold work by a suitable anneal, the relatively small increase in properties will be due to the work performed in the last pass. The interpass anneal should be at a high enough temperature and for a sufficient length of time to obliterate the cold-working stresses and to get rid of the elongated structure thus imparted. Possibly the best method of finishing is to bring to desired gage just before the last pass and finish with a light sink in the final drawing operation. Closer adherence to size would be possible and the finished tube would have a much cleaner and smoother surface.

Steel of about 0.20 to 0.30 per cent. carbon with this method would appear entirely satisfactory for the following physical properties:

Minimum yield point, pounds per square inch..... 36,000
Minimum tensile strength, pounds per square inch..... 55,000
Per cent. elongation in 2 in. { 25 per cent. for 18 B.W.G. and heavier. / 20 per cent. for gages lighter than 18 B.W.G.

The tube to show no cracks when a section 1½ times its outside diameter is compressed until in one zone the same has been increased 25 per cent., or one complete fold is formed.

The tube to show no cracks when flattened until the two parallel walls are no farther distant from each other than twice the thickness of the wall.

Necessity for Flattening Test.—It has been found that in many cases a tube of satisfactory yield point and tensile strength and good ductility, as measured by the compression test, will fail to meet the flattening test. In other instances, a tube of practically the same tensile values will meet both the compression and the flattening tests, the chemical composition in each case being practically the same. In a few photomicrographs, it was noted that the alignment of grain due to cold work was marked in those cases where the tube failed to stand flattening. Without doubt the thermal treatment given in the first case cited was not so carried out as to remove the effect of the heavy reduction on cold passes prior to the last. As these tubes are very often flattened in various degrees, when used as structural materials, a batch having varying characteristics as the above would therefore be unsuited to the purpose intended.

Effect of Size and Gage on Results.—Throughout the foregoing, no mention has been made of the effect of size and gage on the results obtained. Of these two variables, the latter has the more marked effect. Variations in size within reasonable limits do not affect the tensile properties of a given gage tube to any great extent. As the wall thickness decreases, the elongation for a given tensile strength decreases somewhat.

Heat-treated Tubes.—When the desired physical properties cannot be met by either of the methods described, it is better to use a heat-treated tube. The two principal types, as mentioned previously, are 0.30 to 0.40 per cent. carbon and 3½ per cent. nickel with about 0.30 per cent. carbon. Many difficulties are encountered in properly quenching long tubes of small size and thin wall. The principal one is due to the small mass and the tendency for such material to cool very rapidly in air as soon as removed from the heating unit, before the same can be introduced into the quenching bath. It is also difficult to handle such tubes hot, transferring them from the furnace to the bath. Considerable discussion regarding several methods of heating and handling such material can be entered into but is not properly included in this paper. As far as the writer knows, no method that is entirely satisfactory is now in use in America. For short lengths of tube, ordinary methods of heat treatment can be successfully applied.

At the present time, tubes are heat treated after the final cold reduction, which is accompanied by varying degrees of distortion and oxidation of the surface of the finished tube. The writer believes that when satisfactory methods of quenching long thin-wall tubes are in use (and this is not too difficult to accomplish) it will be possible to quench and temper prior to the last cold pass, which may be followed by a bluing.

Results of Varying Conditions of Work and Thermal Treatment on Heat-treated Tubes.—A few results on 0.30 to 0.40 per cent. carbon and 3½ per cent. nickel-steel tubes in varying conditions of work and thermal treatment are given in Tables 9 and 10.

TABLE 9.—*Properties of Hard-drawn and Partly Annealed* 0.30 *to* 0.40 *Per Cent. Carbon Seamless Steel Tubes*

Analysis, Per Cent.				Size and Gage (B.W.G.)	Yield Point, Lb. per Sq. In.	Tensile Strength, Lb. per Sq. In.	Per Cent. Elongation in		Compression at 25 Per Cent.	Flattening Test
Carbon	Manganese	Phosphorus	Sulfur				2 In.	8 In.		
0.31	0.44	0.013	0.026	1 1/16 × 22	91,000	105,000	17.5	9.8	O.K.	O.K.
0.31	0.44	0.013	0.0261	1¼ × 17	62,250	80,750	16.0	9.5	O.K.	O.K.
0.36	0.60	0.016	0.026	1 × 14	67,000	80,900	16.0	9.2	O.K.	Cracked
0.31	0.44	0.013	0.026	¾ × 20	60,000	77,500	13.0	7.0	O.K.	Badly cracked
					65,000	81,000	13.0	8.2	O.K.	Slight cracks
0.31	0.44	0.013	0.026	⅝ × 20	61,000	98,000	20.0	11.7	O.K.	Slight cracks
0.39	0.49	0.018	0.022	1 × 14	64,800	79,550	25.0	12.5	O.K.	Cracked
					63,500	80,000	26.0	14.0	O.K.	Cracked
					60,500	80,900	25.0	12.7	O.K.	Cracked
					65,500	80,100	25.0	12.5	O.K.	Cracked

NOTE.—The drawn tubes annealed at 1000 to 1100° F.

WELDED STEEL TUBES

There are two types of welded steel tubes, those which have been classed as a hot product, where the entire tube is heated to a relatively high temperature when welded, and those in which but a small part of the entire tube is heated in forming the weld, here called a cold product. The methods of producing lap-welded and butt-welded pipe, as the two varieties of the first class are commonly called, have long been known and used. The second type, including acetylene- and electric-welded tubes, is the class of interest in connection with this discussion, being used for non-stressed or low-stressed parts, in various branches of the automotive industry.

The manufacture of such tubing is simpler than that of seamless tubing. Low-carbon (usually under 0.20 per cent.) strip or sheet cut to the desired width and of the proper gage is passed through a forming machine with suitably designed rolls, which gradually bend it into cylindrical form. The tube is then either acetylene or electrically welded. By suitable horizontal rolls the tube is passed under one or more burners,

TABLE 10.—*Properties of Cold-drawn 3½ Per Cent. Nickel-steel Seamless Tubing Under Varying Thermal Treatment*

Tube	Mark	Size and Gage (B.W.G.)		Yield Point, Lb. per Sq. In.	Tensile Strength, Lb. per Sq. In.	Per Cent. Elongation, In.			Compression at 25 per cent.	In flattening tube failed at	Remarks
						2 In.	4 In.	6 In.			
A	1	⅝×18	Anneal approximately 1000° F.	87,230	98,790	19.5	12.5	10.8	sse	Flat	
A	2		1500 to 25° F., oil, 1200° salt, 30 min.	83,630	102,590	23.5	18.0	13.5	sc	Flat	
B	3		As for sample A1	90,970	105,270	15.0	11.0	10.3	O.K.	Flat	
B	4		1500 to 25° F. oil, 1025 to 50° F. salt, 30 min.	121,550	128,240	14.5	10.0	8.3	O.K.	40.4 per cent.	Fine cracks
C	5	1⅛×17	As for sample A1	97,050	112,120	23.5	15.5	....	O.K.	70.7 per cent.	Fine cracks
C	6		1500 to 25° F. oil, 1200° salt, 30 min.	88,980	111,250	27.0	17.0	15.7	O.K.	78.0 per cent.	Fine cracks
D	7		As for sample A1	82,720	101,530	24.0	16.0	12.8	O.K.	84.0 per cent.	Fine cracks
D	8		1500 to 25° F. oil, 1025 to 50° F. salt, 30 min.	112,580	121,250	13.0	9.7	7.7	Cracked	60.4 per cent.	Fine cracks
								8 In.			
		2 ×18	Anneal approximately 1000° F.	76,770	92,730	33.0	24.8	18.6	O.K.		
				100,450	107,890	14.5	12.8	10.4	Cracked		
				92,340	103,050	12.5	10.8	9.0	Fine cracks on		
				107,310	116,440	16.0	12.5	9.2	complete fold		

NOTE.—sse, means slight skin cracks. sc, means skin cracks. 1500 to 1525° F. oil, 1025 to 1050° F. salt, 30 min., means heat to 1500° to 1525° F. and quenched in oil; drawn in salt for 30 min. at 1025 to 1050° F. Analyses for above material are

SIZE	CARBON, PER CENT.	MN, PER CENT.	P, PER CENT.	S, PER CENT.	NI, PER CENT.
⅝×18	0.35	0.45	0.005	0.027	3.42
1⅛×17	0.34	0.58	0.011	0.037	3.39
2 ×18	0.27	0.45	0.016	0.036	3.32

which may be adjusted until brought into the proper position, the flame playing down directly on to the line where the two edges of the strip meet. The speed of travel and the flame are properly coördinated. The rolls are so adjusted that they exert a pressure on the formed strip pressing the edges of the same together. The forming rolls and welding equipment may be located on one stand, the strip entering the rolls at one end and the welded tube coming off the other, or the welding equipment may be located on separate stands, making two operations in the formation of the tube. Electric welding is carried out in a similar manner but by resistance of the gap between the two edges of the formed strip. The weld here formed, however, is not a true butt weld, the edges of the strip being mushroomed by pressure of the contact disks and the rolls.

In either case the resulting burr formed is trimmed off by various methods, such as grinding, strapping, etc. The practice of finishing the tubes from this point varies considerably. In some cases, the tubes are given no further work while often they receive a finishing pass through a die for sizing only. Sometimes they are cold drawn. Probably the most satisfactory tube is produced from a soft strip, which is merely sized by rolls after welding. With welded tubes, considerable variation in tensile properties is encountered and unless the same are annealed and drawn, as for seamless tubes, it is very difficult to obtain a uniform product. It is evident that these properties will vary, depending on the variation in the flame applied and the consequent variation in heating and cooling and amount of metal affected in welding. This variation is present, not only in different tubes but also in individual tubes. Other difficulties are met in electric welding. Unless the welding is carefully controlled, there may be spots throughout the entire length of the tube where no junction has been effected. Welded tubing cannot be cold drawn to anywhere near the same extent that seamless tubing can be worked. In general such tubing cannot compete with seamless tubing but it has its own and important field. High ductility is about all that can justly be required from such a product so that the tube may be bent, flattened, swaged, easily cut and drilled, and otherwise readily handled. Flattening tests with the weld at the point of maximum bend and compression tests are applied and readily show poor welds (see Fig. 6) or lack of ductility.

It is obviously impossible to discuss all the factors entering into the manufacture of products requiring so many varied operations, particularly in a paper of this character. For satisfactory light-wall tubing of uniform physical properties, close control of manufacturing operations is necessary. In addition, a sound understanding by the purchaser is necessary in order that designing engineers will not make their requirements so severe that they cannot be met. Coöperation on the part of both will lower costs and increase production.

DISCUSSION

W. P. PUTNAM, Detroit, Mich.—There is one company that is heating these tubes in an upright position by passing electrical energy directly through the tube itself. The lower connection is attached to an indicator, which shows the point at which the steel passes through the critical range. At that point the clamps are released and the tube is dropped into a well of oil. This method simplifies the heat treatment of tubes in long lengths.

H. J. FRENCH.—I am not entirely familiar with the method outlined, but it appears to have one serious disadvantage; that is, the length of tubing that is directly connected with the electrical contacts is wasted and must be cut off due to either heavy oxidation or burning. That is a serious difficulty, if true in all cases, and cannot be overcome, inasmuch as it is very difficult to draw long tubes of thin wall. Having once obtained a lot of tubes, to cut, say 14 ft. in length, there is often little excess metal available to discard. Am I correct concerning the burning or heavy oxidation of the metal where it comes in contact with the tube?

W. P. PUTNAM.—I have no information regarding the burning; it is, however, necessary to discard the ends of the tube at the point of contact with the electrodes.

H. J. FRENCH.—In one method being developed at the present time these tubes are placed in racks, similar to those used in galvanizing pipe in some plants. These racks, with the tubes, which in effect give greater mass to the material to be treated, are placed vertically in an electrically heated chamber, removed by a crane, and the entire apparatus is immersed in the oil.

Aircraft Steels

BY ALBERT SAUVEUR,* CAMBRIDGE, MASS.

(Chicago Meeting, September, 1919)

As director of the Division of Metallurgy of the Technical Section of the Air Service, American Expeditionary Forces, from August, 1917, to January, 1919, I devoted much time to the study of the steels best suited to the construction of aircraft. It became apparent at a very early stage of my investigation that the practice of builders varied greatly in regard to the number of different steels to be used, to the kinds of steels, and to their treatment. Some manufacturers employed as many as fifteen steels while the builder of one of the newest motors, favorably passed upon in the United States, assured me that two or, at the most, three kinds of steel were all he needed to produce satisfactory results. In the matter of treatment, the manufacturers were also greatly at variance. Some treated their steel with considerable refinement while others paid but scant attention to this feature. Some manufacturers case-hardened many parts, while others used the case-hardening process sparingly, if at all. Nickel-chromium steels were used at some works for case-hardening while at other plants this treatment was applied exclusively to carbon steels. Air-hardening nickel-chromium steel was used extensively by some and not at all by others. Considerable variance existed likewise in regard to the steels used for admission and exhaust valves. The conclusion was obvious, that no two manufacturers of airplane motors held like opinions as to the steels to be used or as to the treatment they should receive. Was it to be inferred from this that the many committees and associations that had been created, in part at least, for the standardization of aircraft steels had failed to make proper recommendations or that the manufacturers had ignored their recommendations?

Confining our inquiry to the committees and associations that had dealt with this question in England, France, and the United States, it was found that in Great Britain the British Engineering Standards Committee, now the British Engineering Standards Association, had devoted much time to the problem of satisfactory standards for aircraft steels. In April, 1918, this Association presented a set of specifications described as "British Standard Specifications for Aircraft Steels for Government Purchases in the United States of America." They included twenty-one different steels and, as the title indicates, were drawn for the

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special purpose of meeting the needs of English aircraft builders and, I venture to add, with but slight regard to what the American steelmakers were prepared to supply. They permitted the use for all purposes of steel containing as much as 0.06 per cent. phosphorus and 0.06 per cent. sulfur and, for a number of reasons, could not have been expected to prove internationally acceptable.

The British Air Ministry had issued a voluminous set of specifications described as "The Air Board Specifications." Not less than seventy of these dealt with aircraft steel. They referred chiefly to English steels and were frequently revised.

In France, the Section Technique de l'Aeronautique Militaire (Ministère de l'Armement) had issued a set of specifications for aircraft steels comprising the description of twenty-two steels, while the "Chambre Syndicale des Constructeurs d'Automobiles," had prepared, at the request of the Ministère de l'Armement, a table of twelve standard steels considered sufficient for the construction of airplane motors.

In the United States, the "International Aircraft Standards Board" was created, I believe, in 1917. It was, however, international only in name, its specifications having been framed chiefly, if not altogether, by the American members to suit American conditions and views, with little if any advice from or consultation with experts of other countries. Not less than fifty specifications dealing with aircraft steel were issued by this Board under the caption "Department of Commerce, Bureau of Standards, International Aircraft Standards adopted by the International Aircraft Standards Board." They were adopted and reprinted with new designations but with slight modifications by the Signal Corps of the United States Army. Not less than thirty-nine different steels were mentioned in these specifications, the system of numbering them being that recommended by the Society of Automotive Engineers.

It must be apparent from the foregoing considerations that if builders of aircrafts still held widely diverging opinions as to the steels they should use and how they should treat them, it was not because of lack of efforts to help them by way of recommendations and specifications on the part of a rather imposing array of commissions, made up of a large number of steel metallurgists and other experts. Rather must it be ascribed to the somewhat bewildering number of standards thrust at them and to the conflicting character of some of them. Not a few builders, moreover, were but incompletely informed regarding the results of the labor of these various bodies. Some hardly knew of their existence; indeed the committees themselves in any one of the three countries mentioned were surprisingly ignorant of the work of similar committees in other countries. I shall even venture the opinion that some of their members were not even interested in what those other committees were doing; that they had no desire to learn of their activities, being well

satisfied with their own methods and progress. We face here a distressing lack of coöperation, of coördination, of interest even in each other's work, necessarily resulting in needless duplications, in waste of time, labor, and money, and in greatly reduced efficiency. Is it to be wondered at if many manufacturers felt somewhat dazed at this deluge of specifications and that, not being able to decide what advices to follow, rejected them all?

It would seem as if there were need of a committee for the Standardization of Standardization Committees, in order to develop method, coöperation, and coördination, without which the result of their labor must remain altogether out of proportion to the energy, time, and money expended.

It is certain that during the entire period of the war no set of specifications for aircraft steels was framed that could in any sense of the word be considered as international, our own "International Air Board" notwithstanding. The specifications of that Board were certainly not recognized as international either in England or in France. Indeed they were hardly known in those countries.

An "Inter-Allied Committee on Aviation" was in existence in 1918 with headquarters in Paris. To one of its sub-committees the study of aircraft steels was entrusted. Several meetings were held, attended by all the members, and the report of the Chairman was adopted with slight modifications, the resolution reading as follows:

"The Sub-committee adopts the present report as being a useful work of classification and of description of the steels used in aviation and as constituting a preliminary step toward the establishment of international specifications. The Committee believes that its future work should include a general study of the testing methods to be applied for the inspection and acceptance of aircraft steels as well as of the methods to be followed in the purchase of steels, between allied countries."

Signed for
France: Commandant Grard,
Commandant Martinot-Lagarde,
England: Captain Neal,
United States: Albert Sauveur,
Italy: Lieutenant Sauda.

Approved by the Permanent Consulting commission, Oct. 26, 1918, for

France: Colonel Dhé,
England: Colonel Peal,
United States: Colonel Dunwoody,
Italy: Commandant Guidoni,

Ratified by the Inter-Allied Committee on Aviation, Oct. 28, 1918. A few copies of this report were published in November, 1918, by the

Information Section, Air Service, American E. F., being inscribed "Technical Section Bulletin, No. 4, Air Service Bulletin." It is here reproduced *in extenso*. It had been hoped to limit the number of steels recommended for the construction of aircrafts to the eleven types described in Category 1, but in order to satisfy all the members of the Committee, additional steels had to be included; by placing them in a separate category, however, their secondary importance has been emphasized.

An International Commission for Aeronautic Standardization was created in 1918. It held its first meeting in London under the auspices of the British Engineering Standards Association on Oct. 14, 15, and 16, 1918. The organization of an American Standards Association is also under way.

Let us hope that these new bodies will profit by the mistakes of their predecessors, that the fruit of their labors may be more commensurate to the energy, time, and money that will have to be expended in again taking up the solution of the same problems apparently left unsolved or but partly solved by earlier organizations.

Report of the Sub-committee on Steels of the Inter-allied Committee on Aviation Adopted 1918

Membership of the Sub-committee

France

Commandant Grard,
Commandant Martinot-Lagarde,
Capitaine Barthelemy,
Lieutenant Sabatier.

Italy

Lieutenant Sauda.

England

Lieutenant-Colonel Jenkin,
Major Thurston,
Captain Neal.

United States

Albert Sauveur, Chairman and Secretary (Rapporteur).

Classification of Aircraft Steels

The steels used in aviation may be divided in two categories:

Category 1 will include all steels very generally used by manufacturers, and which, therefore, may be considered as indispensable.

Category 2 will include steels of very limited or special use or employed by a small number of manufacturers, or which may be used when it is impossible, or at least difficult, to obtain the corresponding steels of category 1; or again, which are not strictly speaking aircraft steels, but which are nevertheless of interest to manufacturers, such, for instance, as tool steels.

These steels will be further grouped in several classes according to their chemical composition: each class will be designated by an appropriate number, the figures of which will have the following signification:

1 will indicate the presence of carbon in the steel,

2 will indicate the presence of nickel in the steel,

3 will indicate the presence of chromium in the steel,
4 will indicate the presence of tungsten in the steel,
5 will indicate the presence of silicon in large proportion.

Class 1, therefore, will correspond to carbon steels; class 12 to steels containing carbon and nickel; class 13 to those containing carbon and chromium; class 123 to those containing carbon, nickel and chromium; class 14 to those containing carbon and tungsten; class 134 to those containing carbon, chromium and tungsten; class 15 to those containing carbon and a large proportion of silicon.

The various steels belonging to the same class will be distinguished from each other by adding to their class number the consecutive letters *a*, *b*, *c*, and so forth, the progression of which will correspond generally to an increasing percentage of carbon. Steel 123-*a*, for instance, will indicate a nickel-chromium steel containing a small proportion of carbon, while steel 123-*b* will be a steel of the same class but more highly carburized. Should it be necessary later to introduce a steel containing a percentage of carbon comprised between two grades already classified, it would then be designated by the use of the two letters of these steels. Steel 123-*ab* for instance would then indicate a steel belonging to class 123 and containing a percentage of carbon included between that of steel 123-*a* and that of steel 123-*b*.

Adopting this classification, the following steels should be considered as interesting the manufacturers of motors and of planes, and their chemical and physical properties should be described:

CATEGORY 1

Class 1.—Carbon steels.

Type 1-*a*.—Extra-soft steel containing from 0.05 to 0.15 per cent. carbon.
Type 1-*b*.—Soft steel containing from 0.15 to 0.25 per cent. carbon.
Type 1-*c*.—Medium-soft steel containing from 0.25 to 0.35 per cent. carbon.
Type 1-*d*.—Medium-hard steel containing from 0.35 to 0.45 per cent. carbon.
Type 1-*e*.—Hard steel containing from 0.60 to 0.80 per cent. carbon.

Class 13.—Chromium steels.

Type 13-*b*.—Chromium steel for ball bearings.

Class 14.—Tungsten steels.

Type 14-*a*.—Steel for magnets.

Class 123.—Nickel-chromium steels.

Type 123-*a*.—Case-hardening steel.
Type 123-*b*.—Medium-hard steel.
Type 123-*c*.—Self-hardening steel.

Class 134.—Chromium-tungsten steels.

Type 134-*a*.—High-speed steel for valves.

CATEGORY 2

Class 1.—Carbon steels.

Type.— 1-*bs* Soft steel high in sulfur.
Type.— 1-*f* Extra-hard steel containing from 0.80 to 1.25 per cent. carbon.

Class 12.—Nickel steels.

Type.— 12-*a* Extra-soft nickel steel.
Type.— 12-*b* Soft steel containing 5 per cent. nickel.
Type.— 12-*c* Steel containing 3½ per cent. nickel.
Type.— 12-*d* Nickel steel non-magnetic.

Class 13.—Chromium steels.

Type.— 13-*a* Steel containing from 12 to 13 per cent. chromium.

Class 15.—Silicon steels.

Type.— 15-*a* Steel for magneto armature sheets.
Type.— 15-*b* Mangano-silicon steel.

Class 1234.—Nickel-chrome-tungsten steel.

Type.—1234–*a* Steel for special gears.

In the description that follows of the chemical and physical properties of aircraft steels, attention should be called to the following points:

While the maximum allowable percentages of sulfur and phosphorus have generally been indicated as 0.05 per cent., this Committee does not desire to place itself on record as holding the view that as much sulfur and phosphorus may be present in high-grade steel without, in the least, impairing its quality. It is not deemed expedient, however, to require lower limits during the war. Similar reservations are made regarding the maximum manganese contents of some of the steels described.

The tensile strength and the yield point have been given in every case both in kilograms per square millimeter and in pounds per square inch, but round numbers rather than exact equivalents have been indicated.

The elongation is given both for a gage length of 100 mm. and of 2 in. It has been assumed that an elongation measured in 10 mm. was equivalent to about seven-tenths of the elongation measured in 2 inches. The accuracy of this relation, however, should be verified.

Impact tests both of the Charpy and Izod types have been given in some cases. It should be understood that the Charpy test refers to the use of the Mesnager test bar, while the Izod test refers to the use of an Izod test bar with a 45° notch.

The relation, Izod = Charpy $\times$ 5.5 $-$ 10, between the Charpy impact test and the Izod test appears to give satisfactory results when applied to steels having a Charpy resilience in the vicinity of 10 kilogrammeters, but it does not seem that the formula can be applied to steels having a much greater or smaller resilience.

Category 1

Steel No. 1-a

Class.—1.

Type.—*a*.

Designation.—1-*a*.

Description.—Extra soft steel.

Adaptability.—For cemented parts subjected to severe frictional wear but not to severe fatigue stresses, also for cylinder sheets.

Examples.—Camshafts, wristpins, tappets, thrust bushings, some rollers and gears, cylinder sheets.

Chemical Composition:

Carbon	0.05–0.15 per cent.
Manganese	not over 0.50 per cent.
Phosphorus	not over 0.05 per cent.
Sulfur	not over 0.05 per cent.

Heat Treatment Recommended.—The parts should be carburized at 900–950° C. (1652–1742° F.) for a length of time depending on the thickness of the case desired. A case 1 mm. thick will generally be produced in 2½ to 3 hr. if a suitable case-hardening mixture be used. The annealing box should be removed from the furnace and the parts allowed to cool in the box, if they are likely to warp; otherwise more quickly in air. They should then be reheated to 900–925° C. (1652–1697° F.) and quenched in water or oil. They should then again be heated to 800° C. (1472° F.) and quenched in water or oil. As a final treatment they may be tempered in oil at 200° C. (392° F.).

Physical Properties.—After the above mentioned heat treatment, the physical properties of the core should be the following:

Tensile strength, min. 55 kg. per sq. mm. or 78,000 lb. per sq. in.

Yield point, min. 32 kg. per sq. mm. or 45,000 lb. per sq. in.

Elongation in 100 mm. 20 per cent.; in 2 in. 28 per cent.

Impact test (Charpy) 22 kg. per sq. cm.; (Izod) 50 ft.-lb. Shore scleroscope, case, min. 75.

Corresponding Specifications of Allied Countries.—France, Section Technique de l'Aeronautique Militaire Specification No. 10.

England, British Standard Specifications for Aircraft Steels No. AL3. British Air Ministry Specification No. SL3.

United States, International Aircraft Standards Board Specification No. 3S1. Signal Corps Specification No. 10030. Standard Steel No. 1010.

Steel No. 1-b

Class.—1.

Type.—*b.*

Designation.—1-*b.*

Description.—Soft steel.

Adaptability.—For parts subjected neither to severe frictional wear nor to high stresses.

Examples.—Cylinder sheets, airplane metal fittings, bolts, screws, pegs, pins, keys, nuts, etc.

Chemical Composition:

Carbon	0.15–0.25 per cent.
Manganese	not over 0.75 per cent.
Phosphorus	not over 0.05 per cent.
Sulfur	not over 0.05 per cent.

Physical Properties.—After heating to 850° C. (1562° F.) and cooling in air, the following physical properties should be obtained:

Tensile strength, min. 42 kg. per sq. mm. or 60,000 lb. per sq. in.

Yield point, min. 26 kg. per sq. mm. or 37,000 lb. per sq. in.

Elongation in 100 mm. 22 per cent.; in 2 in. 30 per cent.

Cold drawn bars: Tensile strength min. 49 kg. per sq. mm. or 70,000 lb. per sq. in.

Yield point, min.

Elongation in 100 mm., 18 per cent.; in 2 in., 25 per cent.

Reduction of area, 45 per cent.

Corresponding Specifications of Allied Countries.—France, Section Technique de l'Aeronautique Militaire Specification No. 12.

England, British Standard Specifications for Aircraft Steels, Nos. AL7 and AL9. British Air Ministry Specifications Nos. S3, S1, part 1.

United States, International Aircraft Standards Board Specifications Nos. 3S26 and 3S20, Signal Corps Specifications Nos. 10028 and 10201; Standard Steel No. 1020.

Steel No. 1-c

Class.—1.

Type.—*c.*

Designation.—1-*c.*

Description.—Medium soft steel.

Adaptability.—For parts subjected neither to severe frictional wear nor to severe fatigue stresses.

Examples.—Bolts, screws, nuts, pins, keys, airplane metal fittings which are not to be soldered.

Chemical Composition:

Carbon	0.25–0.35 per cent.
Manganese	not over 0.75 per cent.
Phosphorus	not over 0.05 per cent.
Sulfur	not over 0.05 per cent.

Physical Properties.—After reheating at 825° C. (1517° F.) followed by cooling in air:

Tensile strength, min. 50 kg. per sq. mm. or 71,150 lb. per sq. in.

Yield point, min., 30 kg. per sq. mm. or 42,690 lb. per sq. in.

Elongation in 100 mm., 20 per cent.; in 2 in., 28 per cent.

Reduction of area, 40 per cent.

Corresponding Specifications of Allied Countries.—France, Section Technique de l'Aeronautique Militaire Specification No. 13.

England, British Standard Specifications for Aircraft Steels, No. AL11. British Air Ministry Specifications Nos. S1 part 2, S6, and S23.

United States, International Aircraft Standards Board Specifications Nos. 3S2 and 3S30, Signal Corps Specifications 10036 and 10251, Standard Steel No. 1030.

Steel No. 1-d

Class.—1.

Type.—*d.*

Designation.—1-*d.*

Description.—Medium hard steel.

Adaptability.—For parts subjected neither to severe frictional wear nor to severe fatigue stresses.

Examples.—Cylinders, magneto-coupling disks, sparking plug caps, spring washers, propeller hubs, bolts, turn-buckles, fork joints, crankcases for rotary motors.

Chemical Composition:

Carbon	0.35 to 0.45 per cent.
Manganese not over	0.75 per cent.
Phosphorus not over	0.05 per cent.
Sulfur not over	0.05 per cent.

Physical Properties.—After heating to 850° C. (1562° F.) followed by cooling in air:

Tensile strength, min. 56 kg. per sq. mm. or 80,000 lb. per sq. in.

Yield point, min. 31 kg. per sq. mm. or 40,000 lb. per sq. in.

Elongation, in 100 mm., 18 per cent.; in 2 in., 25 per cent.

Reduction of area, min. 40 per cent.

Hardness, Brinell, min. 150.

After quenching in oil at 850° C. (1562° F.) followed by tempering at 600° C. (1112° F.):

Tensile strength, min. 67 kg. per sq. mm. or 95,000 lb. per sq. in.

Yield point, min. 42 kg. per sq. mm. or 60,700 lb. per sq. in.

Elongation, in 100 mm., 16 per cent.; in 2 in., 22 per cent.

Reduction of area, 45 per cent.

Hardness, Brinell, 180.

Corresponding Specifications of Allied Countries.—France, Section Technique de l'Aeronautique Militaire Specification No. 14.

England, British Standard Specifications for Aircraft Steels Nos. AL11, AL12, AL13 and AL14. British Air Ministry Specifications Nos. SI part 2, S6, S23, S27, K4 and K5.

United States, International Aircraft Standards Board Specifications Nos. 3S2, 3S30, 3S31, 3S35. Signal Corps Specifications Nos. 10036, 10037, 10251 and 10255. Standard Steel 1040.

Steel No. 1-e

Class.—1.

Type.—*e.*

Designation.—1-*e.*

Description.—Hard steel.

Examples.—Springs, streamlined wires, piano wires.

Chemical Composition:

Carbon	0.60 to 0.80 per cent.
Manganese not over	0.75 per cent.
Phosphorus not over	0.05 per cent.
Sulfur not over	0.05 per cent.

Physical Properties.—After reheating to 780° C. (1436° F.) followed by cooling in the furnace at 600° C. (1112° F.) doors open, and then slowly, the following physical properties should be obtained:

Tensile strength, min. 65 kg. per sq. mm. or 92,500 lb. per sq. in.

Yield point, min. 35 kg. per sq. mm. or 50,000 lb. per sq. in.

Elongation, in 100 mm., 10 per cent.; in 2 in., 14 per cent.

Corresponding Specifications of Allied Countries.—France, Section Technique de l'Aeronautique Militaire Specification No. 15.

Steel No. 13-b

Class.—13.

Type.—*b.*

Designation.—13-*b.*

Description.—Steel for ball bearings.

Adaptability.—Ball-bearing races, valve-rocker rollers.

Chemical Composition:

Carbon	0.90 to 1.10 per cent.
Manganese not over	0.40 per cent.
Phosphorus not over	0.04 per cent.
Sulfur not over	0.04 per cent.
Chromium	1.00 to 1.50 per cent.

Corresponding Specifications of Allied Countries.—France, Section Technique de l'Aeronautique Militaire Specification No. 51.

England, British Air Ministry Specification No. S19.

United States, Standard Steel No. 51100.

Steel No. 14-a

Class.—14.

Type.—*a.*

Designation.—14-*a.*

Description.—Tungsten steel.

Adaptability.—For permanent magnets.

Chemical Composition:

Carbon	0.55 to 0.65 per cent.
Tungsten	5.50 to 6.50 per cent.

Corresponding Specifications of Allied Countries.—France, Section Technique de 'Aeronautique Militaire Specification No. 61.

Steel No. 123-a

Class.—123.

Type.—*a.*

Designation.—123-*a.*

Description.—Nickel-chromium steel for case hardening.

Adaptability.—For cemented parts subjected to severe frictional wear and to severe fatigue stresses.

Examples.—Cams, piston pins, some gears and pinions, pump spindles.

Chemical Composition:

Carbon	0.05 to 0.15 per cent.
Manganese not over	0.50 per cent.
Phosphorus not over	0.05 per cent.
Sulfur not over	0.05 per cent.
Nickel not less than	2.00 per cent.
Chromium not less than	0.50 per cent.
Total nickel and chromium	2.75 to 4.00 per cent.

Thermal Treatment Recommended.—The parts should be carburized at 900–950° C. (1652–1742° F.) for a length of time depending on the thickness of the case desired. A case 1 mm. thick will generally be produced in 2½ to 3 hr. if a satisfactory carburizing mixture be used. The annealing box may then be removed from the furnace and the parts allowed to cool in the box if they are in danger of warping; otherwise more quickly in air. They should then be reheated to 875° C. (1607° F.) and quenched in water or oil; they should then again be heated to 775° C. (1427° F.) and quenched in water or oil. As a final treatment they may be tempered in oil at 200° C. (392° F.).

Physical Properties.—After the above-mentioned heat treatment, the physical properties of the core should be as follows:

Tensile strength, min. 90 kg. per sq. mm. or 128,000 lb. per sq. in.

Yield point, min. 72 kg. per sq. mm. or 102,450 lb. per sq. in.

Elongation, in 100 mm., 12 per cent.; in 2 in., 17 per cent.

Reduction of area, min., 40 per cent.

Impact test (Charpy) 12 kgm. per sq. cm.; (Izod), 25 to 30 ft.-lb. according to tensile strength.

Hardness, scleroscope, carburized case, min. 75.

Corresponding Specifications of Allied Countries.—France, Section Technique de l'Aeronautique Militare Specification No. 31.

England; British Standard Specification for Aircraft Steels No. AL5. British Air Ministry Specification No. S16.

United States, International Aircraft Standards Board Specification No. 3S18. Signal Corps Specification No. 10051. Standard Steel X3315.

Steel No. 123-b

Class.—123.

Type.—*b.*

Designation.—123-*b.*

Description.—Medium-hard nickel-chromium steel.

Adaptability.—For parts subjected to severe fatigue stresses but not to severe wear.

Examples.—Crank-shafts, connecting rods, pump spindles, some pinions, gears, bolts, studs, nuts, etc.

Chemical Composition:

Carbon	0.30 to 0.40 per cent.
Manganese	not over 0.65 per cent.
Phosphorus	not over 0.05 per cent.
Sulfur	not over 0.05 per cent.
Nickel	2.50 to 3.50 per cent.
Chromium	0.50 to 1.00 per cent.

Total nickel and chromium between 3.25 and 4 per cent.

Physical Properties.—After quenching in oil at 825° C. (1517° F.) followed by tempering at 600° C. (1112° F.) the physical properties of the steel should be as follows:

Tensile strength, 88 kg. per sq. mm. or 125,000 lb. per sq. in.

Yield point, 70 kg. per sq. mm. or 100,000 lb. per sq. in.

Elongation in 100 mm., 12 per cent.; in 2 in. 17 per cent.
Reduction of area, min., 45 per cent.
Impact test, (Charpy), 12; (Izod) 40.
Brinell hardness, 269–331.

Corresponding Specifications of Allied Countries.—France, Section Technique de l'Aeronautique Militaire Specifications Nos. 32 and 33.

England, British Standard Specificaions for Aircraft Steels Nos. AL18, AL19, AL20 and AL21. British Air Ministry Specifications Nos. K1, S11, and S12.

United States, International Aircraft Standards Board Specifications 3S4, 3S5, 3S6, 3S33 and 3S34. Signal Corps Specifications Nos. 10046 and 10047, 10048, 10253 and 10254. Standard Steel No. X3335.

Steel No. 123-*c*

Class.—123.
Type.—*c*.
Designation.—123-*c*.
Description.—Self-hardening nickel-chromium steel.
Adaptability.—For parts subjected to heavy loads, fatigue stresses and wear, when a relatively low ductility is permissible.
Examples.—Connecting rods, some shafts, gears, and pinions.

Chemical Composition:

Carbon	0.25–0.40 per cent.
Manganese	not over 0.65 per cent.
Phosphorus	not over 0.05 per cent.
Sulfur	not over 0.05 per cent.
Nickel	3.00–4.50 per cent.
Chromium	0.75–2.00 per cent.

Total carbon, nickel, and chromium not less than 5 per cent.

Physical Properties.—After heating to about 850° C. (1562° F.) followed by cooling in the air, the physical properties should be:
Tensile strength, min. 150 kg. per sq. mm. or 213,450 lb. per sq. in.
Yield point, min. 120 kg. per sq. mm. or 170,000 lb. per sq. in.
Elongation in 100 mm., 6 per cent.; in 2 in., 8 per cent.
Reduction of area, min., 30 per cent.
Impact test (Charpy) min. 6; (Izod) min. 8.

Corresponding Specifications of Allied Countries.—France, Section Technique de l'Aeronautique Militaire Specification No. 34.

England, British Standard Specifications for Aircraft Steels Nos. AL22 and AL23. British Air Ministry Specifications Nos. S18, S28.

United States, International Aircraft Standards Board, Specifications Nos. 3S7 and 3S8. Signal Corps Specifications Nos. 10049 and 10050. Standard Steel No. X3435.

Steel No. 134-*a*

Class.—134.
Type.—*a*.
Designation.—134-*a*.
Description.—High-speed steel.
Adaptability.—For parts subjected to wear at high temperature.
Example.—Valves.

Chemical Composition:

Carbon	0.40 to 0.75 per cent.
Manganese	0.15 to 0.40 per cent.
Phosphorus	not over 0.05 per cent.
Sulfur	not over 0.05 per cent.
Tungsten	14 to 18 per cent.
Chromium	2.50 to 4 per cent.
Vanadium	0 to 1 per cent.

Physical Properties.—To permit machining, this steel should be so treated as to yield a Brinell hardness not exceeding 300.

Corresponding Specifications of Allied Countries.—France, Section Technique de l'Aeronautique Militaire Specification No. 62.

England, British Standard Specification for Aircraft Steels No. AL24. British Air Ministry Specification No. K8.

United States, International Aircraft Standards Board Specification No. 3S38. Signal Corps Specification No. 10040. Standard Steels Nos. W60 and W60-*a*.

CATEGORY 2

Steel No. 1-bs

Class.—1.

Type.—*bs.*

Designation.—1-*bs.*

Description.—Soft steel high in sulfur.

Adaptability.—For parts requiring rapid threading in automatic machines.

Examples.—Screws and nuts.

Chemical Composition:

Carbon	0.15–0.25 per cent.
Manganese	not over 0.75 per cent.
Phosphorus	not over 0.10 per cent.
Sulfur	0.075–0.150 per cent.

Physical Properties.—Tensile strength, 49 to 77 kg. per sq. mm. or 70,000 to 110,000 lb. per sq. in.

Elongation in 100 mm., 11 per cent.; in 2 in., 15 per cent.

Reduction of area, 40 per cent.

Corresponding Specifications of Allied Countries:—England, British Standard Specifications for Aircraft Steels No. AL10.

United States, Standard Steel No. 1114.

Steel No. 1-f

Class.—1

Type.—*f.*

Designation.—1-*f.*

Description.—Very hard steel.

Adaptability.—Tools and some pins.

Chemical Composition:

Carbon	0.80–1.20 per cent.
Manganese	not over 0.65 per cent.
Phosphorus	not over 0.05 per cent.
Sulfur	not over 0.05 per cent.

Physical Properties.—After reheating at 750° C. (1382° F.) followed by cooling in the furnace to 600° C. (1112° F.) doors open, and then slowly, the following physical properties should be obtained:

Tensile strength, min. 75 kg. per sq. mm. or 106,500 lb. per sq. in.

Corresponding Specifications of Allied Countries.—France, Section Technique de l'Aeronautique Militaire Specification No. 16.

England, British Air Ministry Specification No. S25.

United States, Standard Steels Nos. 1080 and 1095.

Steel No. 12-a

Class.—12.

Type.—*a.*

Designation.—12-*a.*

Description.—Extra soft nickel steel.

Adaptability.—Cemented parts subjected to severe frictional wear and to fatigue stresses.

Chemical Composition:

Carbon	0.05 to 0.15 per cent.
Manganese not over	0.50 per cent.
Phosphorus not over	0.05 per cent.
Sulfur not over	0.05 per cent.
Nickel	2.50 to 3.50 per cent.

Thermal Treatment.—The parts should be carburized at 900–950° C. (1652–1742° F.) for a length of time depending on the thickness of the case desired. A case 1 mm. thick will generally be produced in 2½ to 3 hr. if a satisfactory carburizing mixture be used. The annealing box may then be removed from the furnace and the parts allowed to cool in the box if they are in danger of warping; otherwise more quickly in air.

They should then be reheated to 850–900° C. (1562–1652° F.) and quenched in water or oil. They should then again be heated to 775° C. (1427° F.) and quenched in water or oil.

As a final treatment they may be tempered in oil at 200° C. (392° F.).

Physical Properties.—After the above-mentioned thermal treatment, the physical properties of the core should be as follows:

Tensile strength, min. 65 kg. per sq. mm. or 92,500 lb. per sq. in.

Yield point, min. 35 kg. per sq. mm. or 49,500 lb. per sq. in.

Elongation, in 100 mm., 12 per cent.; in 2 in., 17 per cent.

Reduction of area, 45 per cent. minimum.

Impact test (Charpy), 20.

Corresponding Specifications of Allied Countries.—France, Section Technique de l'Aeronautique Miltaire Specification No. 21.

England, British Standard Specifications for Aircraft Steels No. AI4. British Air Ministry Specifications No. S15.

United States: Standard Steel No. 2310.

Steel No. 12-b

Class.—12.

Type.—*b.*

Designation.—12-*b.*

Description.—Soft 5 per cent. nickel steel.

Adaptability.—For parts subjected to heavy loads when a relatively low ductility is permissible, for sheets and cemented parts.

Examples.—Connecting rods for rotary motors, tubes, special axles, sheets, some mented parts.

Chemical Composition:

Carbon	0.05 to 0.15 per cent.
Manganese not over	0.40 per cent.
Phosphorus not over	0.05 per cent.
Sulfur not over	0.05 per cent.
Nickel	4.50 to 6.00 per cent.

Treatment for Cemented Parts.—The parts should be carburized at 900° to 950° C. (1652 to 1742° F.) for a length of time depending on the thickness of the case desired. A case 1 mm. thick will generally be produced in 2½ to 3 hr. if a satisfactory carburizing mixture be used. The annealing box may then be removed from the furnace and the parts allowed to cool in the box if they are in danger of warping; otherwise more quickly in air.

They should then be reheated to 850° C. (1562° F.) and quenched in water or oil. They should then again be heated to 775° C. (1427° F.) and quenched in water or oil.

As a final treatment they may be tempered in oil at 200° C. (392° F.).

Physical Properties.—After the thermal treatment as mentioned above, the physical properties of the core should be as follows:

Tensile strength, min. 105 kg. per sq. mm. or 149,500 lb. per sq. in.

Yield point, min. 78 kg. per sq. mm. or 111,000 lb. per sq. in.

Elongation, in 100 mm., 9 per cent.; in 2 in., 13 per cent.

Reduction of area, 40 per cent.

Resilience (Charpy), 8 (subject to revision); (Izod), 50.

After quenching in oil at 850° C. (1562° F.) followed by re-heating at 600° C. (1112° F.) the following physical properties should be obtained:

Tensile strength, min. 70 kg. per sq. mm. or 99,500 lb. per sq. in.

Yield point, 52 kg. per sq. mm. or 74,000 lb. per sq. in.

Elongation, in 100 mm., 11 per cent.

Reduction of area, 45 per cent.

Corresponding Specifications of Allied Countries.—France, Section Technique de l'Aeronautique Militaire Specification No. 22.

England, British Standard Specification for Aircraft Steels, Specification No. AL6. British Air Ministry Specification No. S17.

Steel No. 12-c

Class.—12.

Type.—c.

Designation.—12-c.

Description.—Medium hard 3½ per cent. nickel steel.

Adaptability.—Medium stressed bolts.

Chemical Composition:

Carbon	0.30 to 0.40 per cent.
Manganese not over	0.75 per cent.
Phosphorus not over	0.06 per cent.
Sulfur not over	0.06 per cent.
Nickel	3.25 to 3.75 per cent.

Physical Properties.—After quenching in oil at 850° C. (1562° F.) followed by tempering at 560° C. (1040° F.):

Tensile strength, min. 70 kg. per sq. mm. or 100,000 lb. per sq. in.

Yield point, min. 49 kg. per sq. mm. or 70,000 lb. per sq. in.

Elongation, in 100 mm., 14 per cent.; in 2 in., 20 per cent.

Reduction of area, 50 per cent.

Impact test (Izod), 40 ft.-lb.

After cooling in air from 850° C. (1562° F.), Brinell hardness, 207 maximum

Corresponding Specifications of Allied Countries.—England, British Standard Specifications for Aircraft Steels, No. AL15. British Air Ministry Specifications Nos. S8, S9, S10.

United States, International Aircraft Standards Board Specifications Nos. 3S3 and 3S32. Signal Corps Specifications 10045 and 10252. Standard Steel No. 2335.

Steel No. 12-d

Class.—12.

Type.—*d.*

Designation.—12-*d.*

Description.—Non-magnetic nickel steel.

Adaptability.—Special bolts, armatures for magnetos.

Chemical Composition.—Nickel, 23 to 25 per cent.

Corresponding Specifications of Allied Countries.—France, Section Technique de l'Aeronautique Militaire Specification No. 23.

Steel No. 13-a

Class.—13.

Type.—*a.*

Designation.—13-*a.*

Description.—12 to 13 per cent. chromium steel.

Adaptability.—For parts having to resist oxidation at high temperature.

Examples.—Valves.

Chemical Composition:

Carbon	0.20 to 0.40 per cent.
Manganese not over	0.50 per cent.
Phosphorus not over	0.035 per cent.
Sulfur not over	0.035 per cent.
Chromium	11.50 to 14.00 per cent.

Physical Properties.—Forged.

Tensile strength, 63 kg. per sq. mm. or 90,000 lb. per sq. in.

Yield point, 48 kg. per sq. mm. or 67,500 lb. per sq. in.

Elongation in 100 mm., 12 per cent., in 2 in., 8 per cent.

Reduction of area, 50 per cent.

Corresponding Specifications of Allied Countries.—England, British Standard Specifications for Aircraft Steels No. AL25, British Air Ministry Specification No. S19.

United States, International Aircraft Standards Board, Specification No. 3S38. Signal Corps Specification No. 10040, Standard Steel No. 51230

Steel No. 15-a

Class.—15.

Type.—*a.*

Designation.—15-*a.*

Description.—Silicon steel.

Adaptability.—For parts requiring a low hysteresis and a very high magnetic permeability.

Examples.—Magneto armature sheets.

Chemical Composition.—Silicon, 3.5 to 4.5 per cent.

Corresponding Specifications of Allied Countries.—France, Section Technique de l'Aeronautique Militaire Specification No. 42.

Steel No. 15-b

Class.—15.
Type.—*b*.
Designation—15-*b*.
Description.—Mangano-silicon steel.
Adaptability.—For parts subjected to twisting and bending stresses.
Examples.—Some axles and springs.
Chemical Composition:

Carbon	0.45 per cent.
Manganese	0.50 per cent.
Phosphorus not over	0.04 per cent.
Sulfur not over	0.04 per cent.
Silicon	1.80 per cent.

Physical Properties.—After quenching in oil at 900° C. (1652° F.) followed by tempering at 500° C. (932° F.):
Tensile strength, 120 to 140 kg. per sq. mm. or 171,000 to 199,000 lb. per sq. in.
Yield point min., 90 kg. per sq. mm. or 128,000 lb. per sq. in.
Elongation in 100 mm., 6 per cent., in 2 in., 8 per cent.
Impact test (Charpy), min. 5 kg. per sq. cm.
Corresponding Specifications of Allied Countries.—France, Section Technique de l'Aeronautique Militaire Specification No. 41.

Steel No. 1234-a

Class.—1234.
Type.—*a*.
Designation.—1234-*a*.
Description.—Nickel-chromium-tungsten steel.
Adaptability.—For parts which should retain considerable hardness at high temperature.
Examples.—Admission and exhaust valves.
Chemical Composition:

Carbon	0.20 to 0.30 per cent.
Manganese not over	0.30 per cent.
Phosphorus not over	0.04 per cent.
Sulfur not over	0.04 per cent.
Nickel not over	4.50 per cent.
Chromium not over	0.30 per cent.
Tungsten not over	2.40 per cent.

Physical Properties.—After quenching in the air at 900° C. (1652° F.):
Tensile strength, 180 to 220 kg. per sq. mm. or 256,000 to 313,000 lb. per sq. in.
Yield point, 165 to 195 kg. per sq. mm. or 234,800 to 277,500 lb.
Elongation in 100 mm., 5 per cent., in 2 in., 7 per cent.
Impact test (Charpy), min. 4 kg. m. per sq. in.
Corresponding Specifications of Allied Countries.—France, Section Technique de l'Aeronautique Militaire Specification No. 71.

DISCUSSION

GEORGE K. BURGESS,* Washington, D. C. (written discussion†).—Professor Sauveur refers to the International Aircraft Standards Board in terms that would indicate his non-familiarity with the organization

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†Received Oct. 6, 1919.

and the working of this Board which, of course, is excusable on account of Professor Sauveur's absence in Europe at the time. He implies that the work of this Board was entirely dominated by American interests, whereas diametrically the opposite was the case. The Board was organized for the purpose of facilitating purchase in the United States, on the basis of common specifications, by the Allied air services in the United States, and the Board performed a very valuable service in unifying the specifications of Italy, Great Britain, France, Canada, and the United States for such purposes. The proceedings of the Board were carried out on the basis of unanimous agreement of the representatives of the five countries and all the members from each of the countries were technical specialists or had the advice of technical assistants. It would be difficult to find an instance of greater unity of purpose, readiness to compromise, and efficient production than was shown by this Board, which produced seventy-eight specifications, in the preparation of which representative American manufacturers appeared before the Board. In no case was a specification adopted until assurance was given not only of adequate manufacturing facilities but of willingness to manufacture under the specification in question. Many of these specifications were taken over bodily by the United States Signal Corps, and they have since been widely quoted in the technical press, handbooks, etc. The work of this Board ceased with the International Aircraft meeting in London in 1918, as it was expected that this later organization would continue the work on a more comprehensive basis for purchases in all allied countries.

Prevention of Columnar Crystallization by Rotation During Solidification

BY HENRY M. HOWE,* LL.D., SC.D., AND E. C. GROESBECK,† S.B., M.A., BEDFORD HILLS, N. Y.

(New York Meeting, February, 1919)

(A CONTRIBUTION FROM GREEN PEACE LABORATORY)

THAT the quiescence of a liquid while it is solidifying should favor the formation of columnar crystals, normal of the cooling surface, is seen readily on considering the mechanism of solidification.

First, each particle of any composite liquid, whether it be an aqueous solution or a molten metal, in solidifying splits up into two parts, different in composition and hence in fusibility. One part is infusible at the existing temperature, and hence solidifies, and in general attaches itself to the inclosing walls of metal which have already solidified; *A*, Fig. 1. The other part is fusible at the existing temperature and hence remains molten. In the case of carbon steel, the part of each drop which actually solidifies is poorer in carbon than the drop itself was before it began to solidify, and this impoverishment of the solidifying half-drop enriches the other half-drop in carbon, and thus makes it unfreezable at the existing temperature.

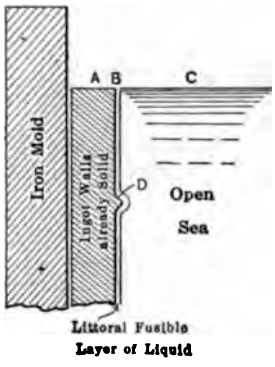

FIG. 1.—MECHANISM OF SOLIDIFICATION.

By this mechanism there arises during solidification a "littoral" or shore layer of liquid *B*, Fig. 1, bathing the already solidified walls, and more fusible than either those walls or the great remaining central mass of liquid or "deep sea" *C*, from which it separates them. It is essential that we grasp clearly this conception of a fusible littoral molten layer coating the already solidified walls and separating them from the deep sea.

Meanwhile heat is flowing rapidly outward through these walls, its escape cooling them, so that if any given particle of the deep sea metal could get past this littoral layer and come into contact with the solid walls, it would in turn solidify, and like its predecessors would split up in solidifying into a less fusible half-drop which would attach itself to those walls and a more fusible one which would remain molten. Thus we see

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that solidification can be continued only by some process of diffusion or convection, which will bring the freezable, because less fusible, deep sea metal past this fusible envelope which coats the solid walls, and into contact with them.

And this brings us to the columnar mode of crystallizing. Any projection, such as *D*, from the face of the walls will be reached earlier than the adjoining smooth unprojecting parts by the freezable deep sea particles. Further, each such projection increases its advantage over the neighboring smooth faces with every fresh addition to its tip. Moreover, this increment of advantage is continuous, and indeed continuously self-exaggerating. Growth at the tips of these projections in itself is the columnar growth.

This sketch of the growth has been somewhat simplified by dwelling on convection as leading freezable droplets past the fusible and unfreezable littoral layer of the molten mass. But what we have sketched as true of convection should be true of diffusion, which is probably more effective than convection in feeding the tips of the protruding crystals.

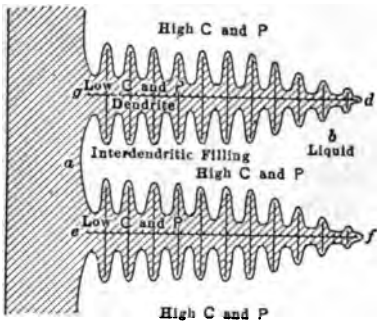

FIG. 2.—CONCENTRATION OF CARBON AND PHOSPHORUS IN THE FILLING BETWEEN ADJOINING COLUMNAR CRYSTALS.

Having come thus far, we take the next step of recognizing that agitation, in that it tends to bring fresh freezable metal from the deep sea to the sides also of the growing crystal, thus diminishes the advantage which the tips have over the sides, and thus diminishes the columnarity of the growth. In particular, if we can keep the molten metal constantly sweeping past our supposed projections, it will in the first place sweep away the fusible littoral layer, and in the second place bring fresh freezable deep sea metal to the sides as well as to the ends of the columns.

The harmfulness of the columnar structure, especially if the columns themselves are coarse, is seen by considering a later stage of this growth, as sketched in Fig. 2. In the case of steel ingots, the splitting up which accompanied the solidification of each droplet rejects not only carbon but also phosphorus into the littoral layer. If the parts of this which wash the sides of the columns are well landlocked, as in Fig. 2, this local concentration of the littoral layer in these elements will continue, and will progressively exaggerate itself as solidification progresses, leading at last to a marked local enrichment or segregation of these and other elements in the part midway between each pair of neighboring crystals. This is very harmful, because this segregate forms a brittle link in the chain where rupture may occur under a shock which would fail to break the metal if it were uniformly distributed.

Continuously varying rotation during solidification should evidently be a very effective way of keeping the molten metal sweeping past the growing walls. This was proposed by Tchernoff[1] before 1880, and was carried out by Webb.[2] Tchernoff pointed out that the rate of rotation for this purpose should be as great as possible at first and that its direction should be changed rapidly and violently.

Webb[2] cast locomotive driving wheels thus, feeding the mold in the center, while rotating it slowly at first, accelerating the rotation to about 40 to 50 revolutions per minute when the mold was filled, and then reducing the speed gradually. These two inventors almost certainly intended

FIG. 3.—UPPER SURFACE OF ROTATED INGOT, SHOWING STRONG SPIRAL MARKINGS. × 1.

to prevent columnar crystallization by this means, though the mechanism of solidification was not known then. Our experiments carry out their forgotten and re-invented process.

In order to cause the continuous variation of rate of rotation, we set the vessel in which solidification is to occur on the horizontal disk of a common polishing machine, such as is used for preparing sections for microscopic examination. By holding the belt which drives this disk in the hands one can readily bring about a very rapid and continuous variation in the rate of rotation, by starting the rotation slowly to the right, clockwise, bringing it rapidly to a maximum and again retarding

[1] D. K. Tchernoff: Structure des Lingots Coulés en Acier. *Rev. Universelle* [2] (1880) **7**, 154.

[2] *Jnl.* Iron and Steel Inst. (No. 2, 1882), 522.

it and replacing it with contra-clockwise rotation. In order to show clearly the effects of this rotation, two like lots of the liquid or molten mass in each experiment were poured in immediate succession and in like manner into two like vessels, one of which was allowed to cool undisturbed while the other was rotated as just described. For brevity this method may be referred to as simply "rotation," which in every case varied continuously.

Experimental Results.—A strong hot solution of ammonia alum gave very much coarser crystals with quiescent solidification than with rotating solidification. Like results were reached with commercial zinc.

In order to introduce still greater differentiation during solidifying, we next tried a mixture of zinc with about 5 per cent. of type metal,

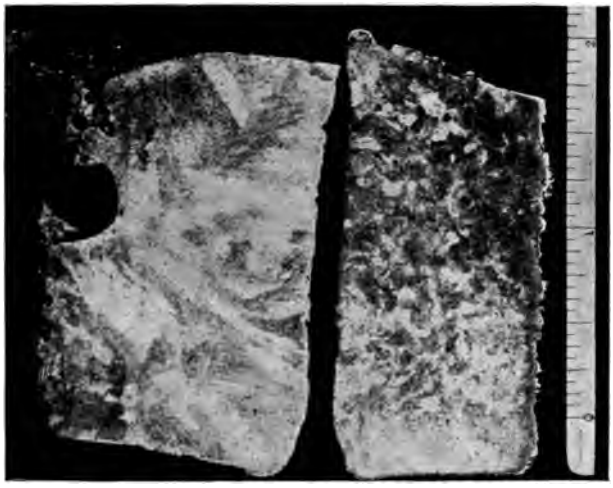

FIG 4. FIG. 5.

FIG. 4.—STATIONARY INGOT. × 1. FIG. 5.—ROTATED INGOT. × 1.
ABOUT HALF OF THE ETCHED VERTICAL SECTION OF EACH INGOT IS SHOWN HERE.

under the following conditions. The metal was melted in a clay crucible under charcoal, was thoroughly stirred, and was then poured half into a stationary iron mold and half into a like mold which was placed on the polishing disk immediately after pouring and started rotating reversingly and kept rotating for a long time after the upper crust had solidified. The conditions were as follows:

	MIN.	SEC.
Poured into stationary mold at	0	0
Poured into rotating mold at	0	15
The rotation was started at	0	50
The upper surface of ingot in the rotating mold was completely frozen over at	6	55
The rotation was continued till	36	35

The strong spiral markings on the upper surface of the rotating ingot are seen in Fig. 3. Note how much more marked the columnar crystals are in the stationary Fig. 4, than in the rotated one, Fig. 5.

In every case in which the rotation was continued for a long time, results like these were reached. In one case in which the rotation was stopped soon after the upper surface had frozen over, no very marked difference between the stationary and rotated ingot appeared. This, we believe, was because most of the solidification occurred after the rotation ceased.

Fig. 6 shows the marked furrows in the central pipe, to the occurrence of which in steel one of us has already called attention.[3]

FIG. 6.—FURROWS IN WALLS OF PIPE IN STATIONARY INGOT. × 1.6.

DISCUSSION

LEWIS B. LINDEMUTH, New York, N. Y.—I would like to confirm Doctor Howe's supposition that a process could be worked out for steel along similar lines. About 1½ years ago, F. D. Carney and myself were working on the elimination of columnar crystals which are particularly evident in nickel steel. We discussed rotation and varying speeds of rotation, but figured that with just rotation the steel would eventually catch up to the motion of the mold and that the desired effect could not be obtained in that way, so we decided to alternate the direction of rotation. We rigged up an ordinary commercial turntable and on the same mounting attached a blast-furnace and mud gun, which is a simple steam cylinder, and mounted an 18-in. ingot mold on the turntable. The turntable had a stroke of about 10° to 12°, with stops at the ends. The steam cylinder would bang the table against the stops so that the stopping was so violent that it would tend to keep the metal in motion when the mold stopped and thus wash away any crystals that would form. We made quite a few interesting experiments; the results of which were that the columnar crystallization was entirely eliminated. At the same time we cast ingots in a similar mold, allowing the molds to remain stationary; these showed columnar crystals of from 2 to 4 in. in depth. The grain structure of the balance of the ingots, which were rotated, was very much finer than the grain structure of the ingots that were cast and allowed to cool slowly. The experiments were made on a 0.35 per cent. carbon nickel steel, which is probably more susceptible to columnar

[3] H. M. Howe: *Trans.* (1907) **38**, 3–108.

crystallization than any other. I think that the rolling properties and the forging properties of some alloy steels could be very much improved by a similar method, or some commercial method of casting ingots to eliminate this columnar crystallization.

T. S. WOODWARD,* Youngstown, Ohio.—I should like to ask either Doctor Howe or the last speaker, what effect this treatment has on axial segregation, the drawing of the impurities, or the nickel in the case of the last experiment, toward the center of the ingot.

LEWIS B. LINDEMUTH.—I cannot give any information as to what the final segregation would be. Because of an accident we obtained some very interesting results. About 4 or 5 min. after the machine had started one of the bolts broke, so it was necessary to stop the machine and repair it, which took probably a minute. The machine was started again and run for 5 min. longer. After the ingot was broken, it showed a distinct line of coarse crystals about ¼ in. wide through its entire length from the bottom to the top. We analyzed this thin section, which showed the coarse crystallization, and the section from the center of the ingot. The center of the ingot showed no segregation at all, but the large crystals showed 0.50 per cent. carbon and about 0.030 per cent. segregation of sulfur and about 0.020 per cent. segregation of phosphorus.

* Chief Chemist, Ohio Works, Carnegie Steel Co. R 2.

Differential Crystallization in a Cast-steel Runner*

BY FRANCIS B. FOLEY,† WASHINGTON, D. C.

(Chicago Meeting, September, 1919)

IN examining steel under the microscope, one is constantly confronted with structures that are difficult to interpret. Recently, in a collection of samples for exhibition purposes, the writer found a piece of open-hearth runner metal, 2 in. (50 mm.) in diameter, that had been fractured for use as a sample of cast-steel structure. On polishing and etching a section for the purpose of photographing a field to typify the microstructure of steel in the "as cast" condition, a variety of structures was found (Fig. 1) so arranged as to arouse interest at once. For a distance of about 0.065 in. (1.6 mm.) from the skin of the runner, normal ingot structure was found (Fig. 2); this was followed by a zone, about 0.215 in. (5.5 mm.) wide, of considerably lower carbon content in which occurred a most pronounced Widmannstättian structure (Fig. 3); from this point on to the center of the runner, the metal was again of normal ingot structure (Fig. 4).

Representative samples taken from the metal in the three zones showed the following analysis:

	CARBON PER CENT.	NICKEL PER CENT.	CHROMIUM PER CENT.
Metal from skin	0.34	1.26	0.63
Widmannstättian zone	0.21	1.18	0.63
Metal from center	0.34	1.32	0.67

Such a condition might result from decarburization followed by recarburization of the skin. This section came from one of the branch runners to the bottom of the mold, not from the main runner, and it is difficult to conceive of decarburization occurring in this portion of a runner. If the metal were stripped while quite hot, some oxidation would take place but the recarburization is unaccounted for. It could not occur by segregation after solidification—the tendency then is for diffusion and homogeneity. The following seems to fit the conditions and is offered as an explanation.

The outside layer of normal structure 0.065 in. (1.6 mm.) thick was chilled on the walls of the runner brick when the molten metal first entered the runner. As the molten metal flows by, the outside of the stream

* Published by permission of the Director of the U. S. Bureau of Mines.

† Metallurgist, U. S. Bureau of Mines.

is cooled by the walls of the runner and crystallization starts; the crystals so formed become attached to the chilled metal on the wall of the runner and, being the first to form, are necessarily of the lowest carbon con-

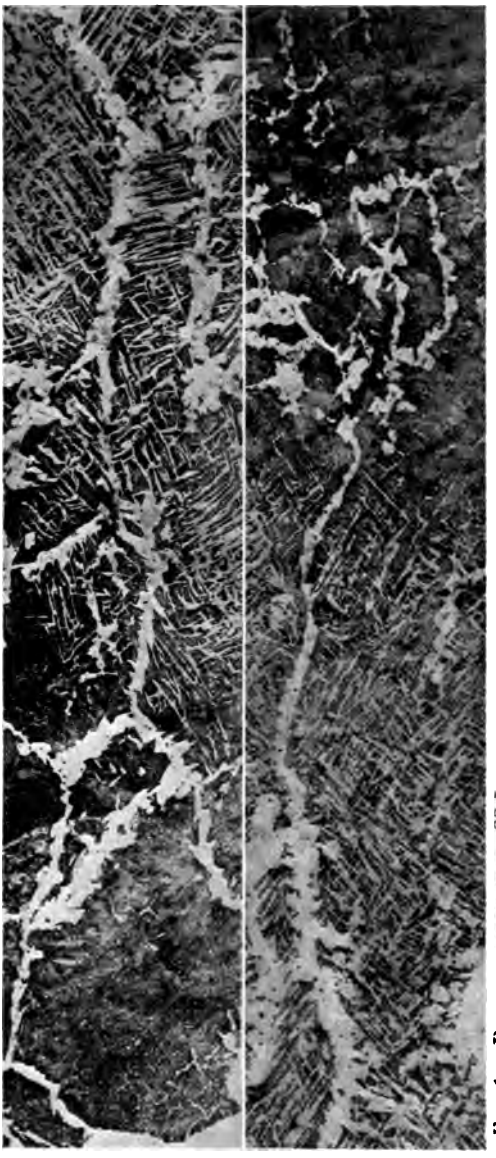

FIG. 1.—PHOTOMICROGRAPH—33.5 DIAM. PICRIC-ACID ETCHING, SHOWING RANGE OF STRUCTURE FROM OUTSIDE TOWARD CENTER OF RUNNER.

centration. When the wall of the runner has reached a certain temperature, the outside of the flowing stream of metal comes to a fairly constant temperature; but as that temperature is just under the liqui-

dus, solid metal of low carbon content is constantly separating out and crystallizing on the walls of the runner. These crystals build outward into the stream. When the difference in temperature between the solid metal attached to the walls of the runner and the molten stream has become so small that no chilling effect is exerted, the formation of these low-carbon crystals ceases, or perhaps they grow until such a length that they

FIG. 2.—100 DIAM. PICRIC-ACID ETCHING STRUCTURE OF EXTERIOR METAL IN RUNNER. ANALYSIS SHOWS C 0.34 PER CENT.

are no longer strong enough to withstand the force of the flowing metal. After the mold has been filled and the metal comes to a state of rest, solidification proceeds normally and the structure of the interior of the runner is the result. In subsequent cooling in the solid state some diffusion takes place, but evidently it does not become complete. This diffusion is first materially retarded when the metal of lowest carbon content passes through its Ar_3 point and is stopped completely at Ar_1.

The very marked Widmannstättian structure in the low-carbon band might have been produced in either of two ways, or by a combination of the two. One possibility is that it is due to the fact that the metal, where this type of structure is so pronounced, existed in the solid state at a higher temperature and for a longer time than the metal elsewhere. Its freezing point is higher than that of either the skin or the center metal,

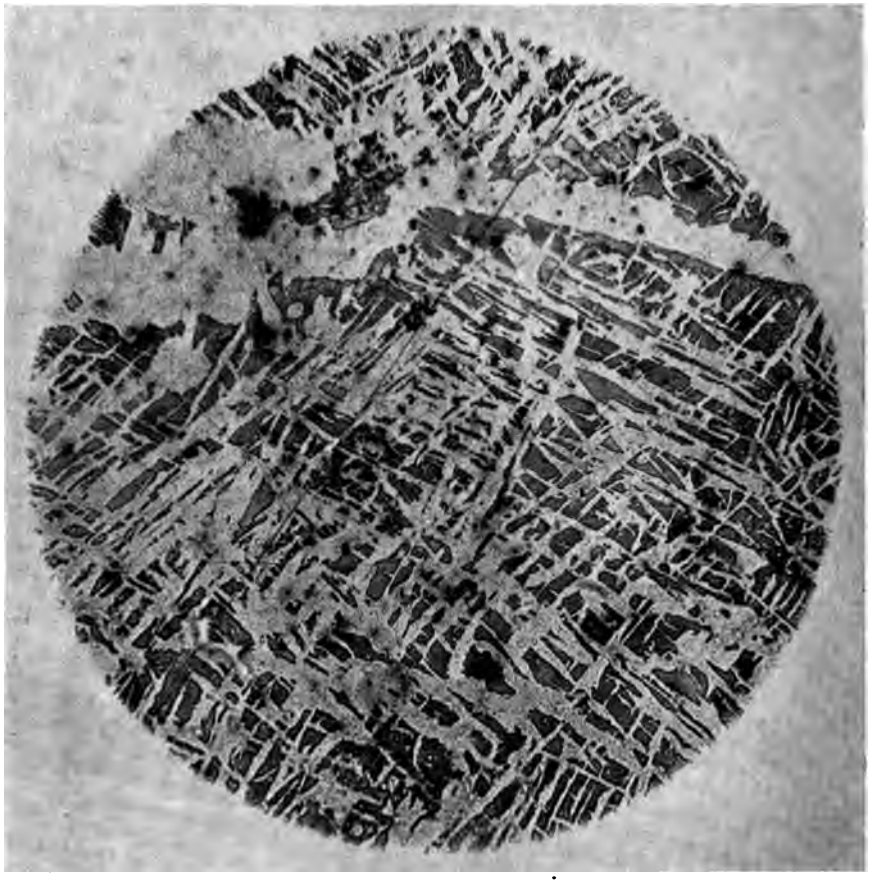

FIG. 3.—100 DIAM. PICRIC-ACID ETCHING. TYPICAL OF WIDMANSTÄTTIAN ZONE IN RUNNER. ANALYSIS SHOWS C 0.21 PER CENT.

and it did not cool as rapidly as the skin but somewhat faster than the center. The other possibility is that this type of structure formed more readily in the particular part of the runner than elsewhere because of the difference in the processes of crystallization, and that the ferrite needles are actually in the same position that the original low-carbon crystals occupied during their formation and growth.

DISCUSSION

HENRY M. HOWE, Bedford Hills, N. Y. (written discussion*).—Mr. Foley's explanation of this interesting occurrence is perfectly reasonable. Possibly a word explaining the phenomenon more in detail may not be amiss.

In the act of solidification each drop as it starts to solidify splits up into a half drop poorer in carbon and another half drop richer in carbon

FIG. 4.—100 DIAM. PICRIC-ACID ETCHING. STRUCTURE OF INTERIOR METAL. ANALYSIS SHOWS C 0.34 PER CENT.

than the drop itself. The half drop poorer in carbon solidifies and attaches itself to the already solid walls, whereas the half drop richer in carbon remains molten for the excellent reason that it is richer and hence is more fusible than the original drop, and hence in turn is wholly fusible at the existing temperature.

* Received Aug. 8, 1919.

Under usual conditions, as when the steel solidifies in a mold of any form, the successive molten half drops thus enriched in carbon remain present and by their enrichment progressively raise the carbon content of the remaining molten mass, out of which successive half drops are continuously solidifying and attaching themselves to the already solid walls. Because the molten mass thus becomes progressively richer in carbon, each of the half drops which solidify from it in like manner is richer than the preceding one, because the carbon content of the particles which thus solidify increases with that of the molten mass out of which they solidify. Hence the familiar segregation, or progressive enrichment of carbon content from the skin or first freezing part of the mass to the matter just below the bottom of the pipe, which is the last freezing part.

But in the case of a runner this last condition is absent. Each molten half drop enriched in carbon and set free by the solidification of its mate poorer in carbon is immediately swept away by the swift stream gushing through the runner, so that every drop which starts to solidify is of the carbon content of the original molten mass in the ladle, and thus the progressive enrichment from drop to drop does not occur. Hence the existence of the low-carbon Widmannstättian band.

That the very outer band of all is much richer in carbon than this one is evidently due to a solidification of the very outer crust of the runner so rapidly, and in the form of pine trees out-shooting so quickly as to entangle and retain their molten higher-carbon mates formed by the act of solidification.

This principle which Mr. Foley points out is new, I believe, and may play a very important part in the phenomena of segregation in general. Thus it helps to explain the impoverishment of the lower deep-seated part of the ingot, from the axis outward for a considerable distance toward the skin. While this part is solidifying there is a rapidly rising molten current along the already solid walls, rising partly because its enrichment in carbon by the act of solidification makes it lighter than the average of the molten mass, and partly because of the entraining action of the gases which are now escaping from solution in effervescing steel. In fact, the formation of the Widmannstättian band may be grouped with this "negative segregation," so called, and so often noticed in the lower peri-axial part of steel ingots. It is segregation, though of a special type.

Heat Treatment of Cast Steel

BY JOHN H. HALL,* ARVID E. NISSEN,† AND KNOX TAYLOR,‡ HIGH BRIDGE, N. J.

(Chicago Meeting, September, 1919)

SOME months ago one of the authors was asked to write a paper on the heat treatment of steel castings that would be more comprehensive than other matter he had published; this is an attempt to present in condensed form the results of experimental work and actual practice carried out for the most part in 1907, 1908, and 1909. It is not a laboratory research and the work done was carried out for the express purpose of putting in practice the processes discovered as fast as possible. Most of the tests are the result of commercial practice and not of bars annealed in our laboratory; most of the micrographs were made for our own records and not for publication. Certain of the micrographs, therefore, are not as clear as we may wish but they were good enough for our purpose at the time they were made. The tests and micrographs are selected from several thousands in our files. All of these micrographs are magnified 60 diameters.

These experiments were conducted almost entirely on steel made in a 3-ton bottom-blown converter, and most of the steel illustrated analyzes about 0.05 in phosphorus and sulfur. Unfortunately, a certain number of specimens are shown of which we do not have the complete analysis, but in most cases in which the phosphorus and sulfur vary much from the figure just given that fact is noted. For purposes of comparison, we have included a certain number of micrographs and tests of steel made in the electric furnace, some of which were from specimens made in other shops and some in our own shop, and some tests of acid open-hearth and tropenas steel.

A brief examination of the tests in Table 1 shows that certain treatments of, for instance, 0.30 per cent. carbon steel are not represented by any test data. Had this been a laboratory research, this information would undoubtedly have been furnished, but as this work was done for commercial purposes the omissions are due to the fact that, at the time the work was done, we saw no advantage in trying every composition of steel with each treatment discussed in the paper.

* Metallurgist, Taylor-Wharton Iron & Steel Co.
† Assistant Metallurgist, Taylor-Wharton Iron & Steel Co.
‡ President, Taylor-Wharton Iron & Steel Co.

Annealing Temperature and Time

When we started to investigate the annealing of cast steel, we thought that we should use a lower annealing temperature for the high-carbon steel than for the low-carbon steel, and we took it for granted that the ingot structure of a heavy casting would take longer to eliminate than the similar structure of a small casting. We accordingly prepared a ½ in. (1.27-cm.) square bar and a 6-in. square cast billet of 0.25 per cent. carbon steel, a similar pair of 0.40 per cent. carbon steel, and of 0.55 per cent. carbon steel. The ½-in. square bars were cut into pieces about ½-in. long, and we cut from the outside and from the center of the 6-in. square billets small pieces about ½ in. cube. These pieces were fastened together with wire, nine pieces being heated at one time. The steel was heated in a laboratory electric-resistance furnace to different temperatures for different lengths of time, and the bundles of pieces were removed from the furnace and cooled in a box of lime, on the supposition that these small pieces would cool slowly enough to be representative of commercial annealing practice, but when we compared the microstructure of steel cooled slowly in a commercial annealing furnace with the microstructure of these small bars we found an entirely different structure. As a matter of fact, we soon came to the conclusion that the small pieces cooled in lime had cooled at about the same rate as a medium heavy casting taken from the furnace and cooled in the air. Figs. 1, 10, and 19 show the structures of the untreated steel of 0.25, 0.40 and 0.55 per cent. carbon, respectively; Figs. 4, 13, and 22 show the outside of the 6-in. square billets; and Figs. 7, 16, and 25 show the center of the 6-in. square billets. All of these represent the steel before any treatment had been applied. As already mentioned, it seemed probable that the large masses of ferrite in Fig. 7, for instance, would be more difficult to obliterate than the small ferrite areas in Fig. 1.

Specimens of each size and kind were heated to 800° C. for ½ hr., 1 hr., and 2 hr.; also to 900° C. and 950° C. for the same lengths of time. All of these specimens were polished and photographed. From the ninety-nine photographs, the twenty-seven selected are shown in Figs. 1–27, inclusive. Examination of these photographs shows that in practically all cases a certain amount of free ferrite remains in the microstructure after heating to 800° C. for 2 hr., and that practically none remains after heating to 900° C. for 2 hr. It is very surprising that in many cases the ½-in. square bars show more free ferrite after annealing for 2 hr. at 800° C. than the outside of the 6-in. square bars, and that apparently the free ferrite of the ½-in. square bars is absorbed at about the same rate as the much larger masses of ferrite at the center of the 6-in. square bars.

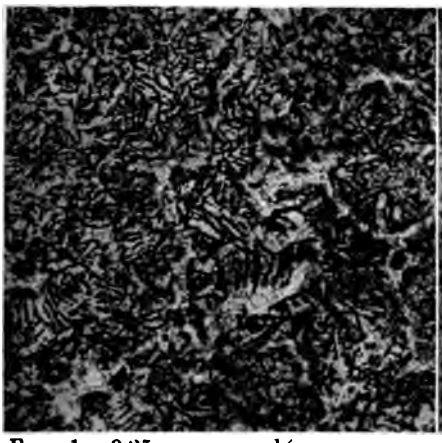

FIG. 1.—0.25 CARBON. ½-IN. SQUARE BAR AS CAST.

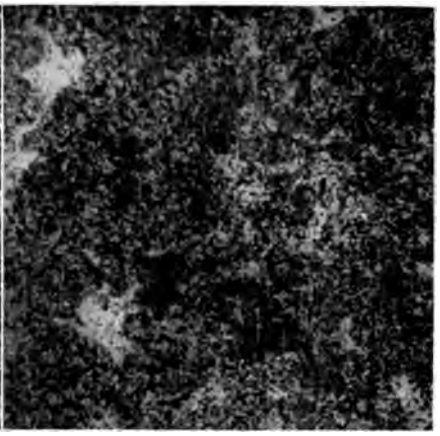

FIG. 2.—SAME HEATED TO 800° C. FOR 2 HR. AND COOLED IN LIME.

FIG. 3.—SAME HEATED TO 900° C. FOR 2 HR. AND COOLED IN LIME.

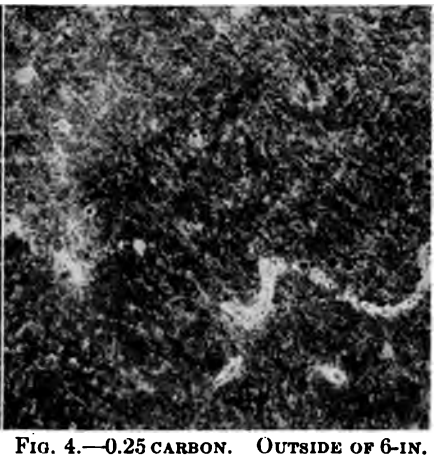

FIG. 4.—0.25 CARBON. OUTSIDE OF 6-IN. SQUARE BAR AS CAST.

FIG. 5.—SAME HEATED TO 800° C. FOR 2 HR. AND COOLED IN LIME.

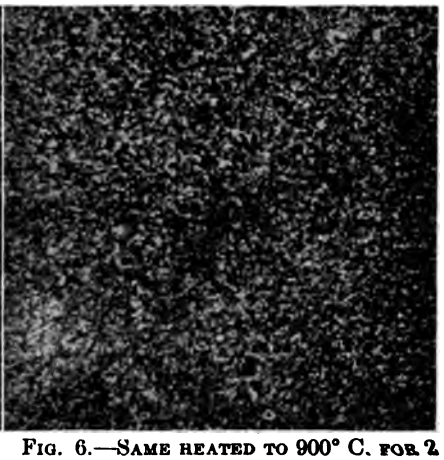

FIG. 6.—SAME HEATED TO 900° C. FOR 2 HR. AND COOLED IN LIME.

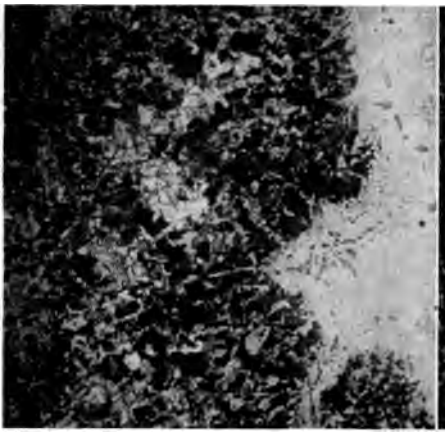

FIG. 7.—0.25 CARBON. CENTER OF 6-IN. SQUARE BAR AS CAST.

FIG. 8.—SAME HEATED TO 800° C. 2 HR. AND COOLED IN LIME.

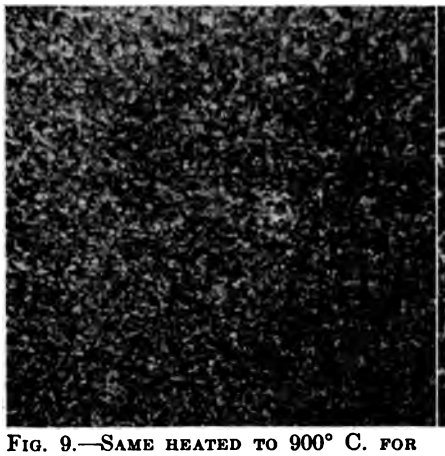

FIG. 9.—SAME HEATED TO 900° C. FOR 2 HR. AND COOLED IN LIME.

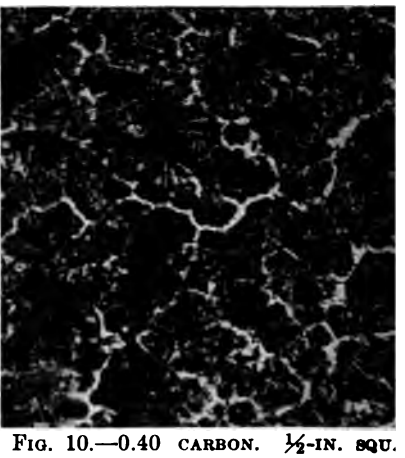

FIG. 10.—0.40 CARBON. ½-IN. SQU. BAR AS CAST.

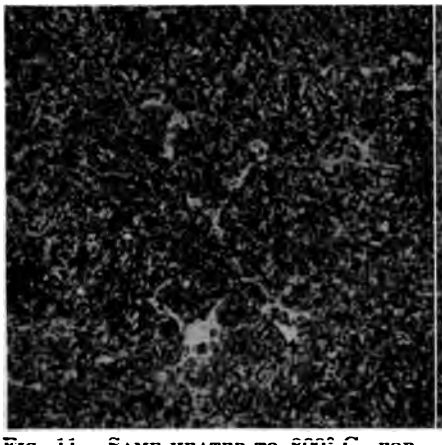

FIG. 11.—SAME HEATED TO 800° C. FOR 2 HR. AND COOLED IN LIME.

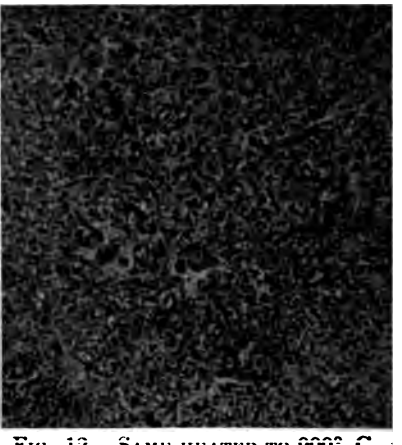

FIG. 12.—SAME HEATED TO 900° C. 2 HR. AND COOLED IN LIME.

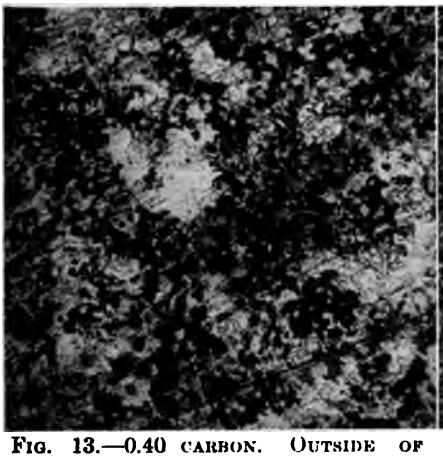

FIG. 13.—0.40 CARBON. OUTSIDE OF 6-IN. SQUARE BAR AS CAST.

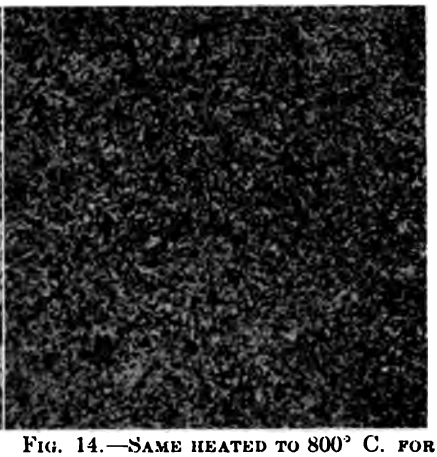

FIG. 14.—SAME HEATED TO 800° C. FOR 2 HR. AND COOLED IN LIME.

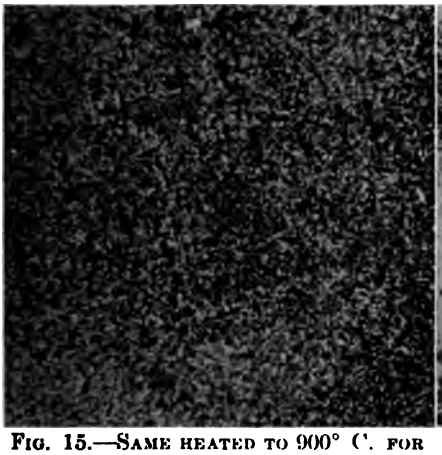

FIG. 15.—SAME HEATED TO 900° C. FOR 2 HR. AND COOLED IN LIME.

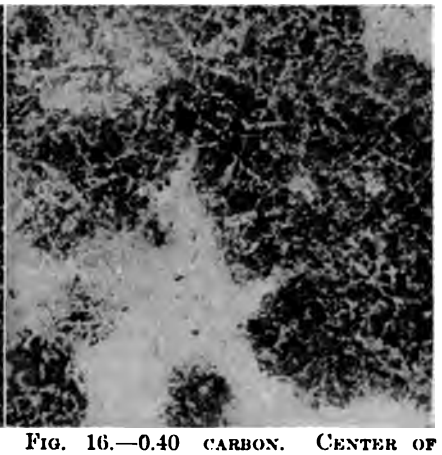

FIG. 16.—0.40 CARBON. CENTER OF 6-IN. SQUARE BAR AS CAST.

FIG. 17.—SAME HEATED TO 800° C. FOR 2 HR. AND COOLED IN LIME.

FIG. 18.—SAME HEATED TO 900° C. FOR 2 HR. AND COOLED IN LIME.

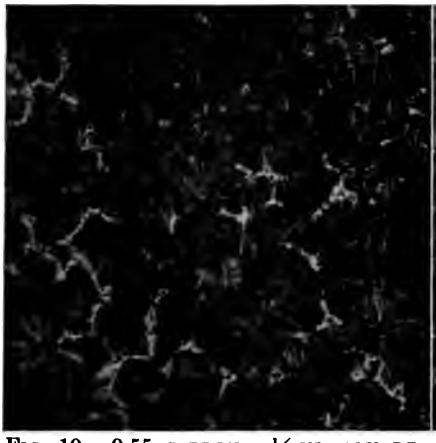

FIG. 19.—0.55 CARBON. ½-IN. SQUARE BAR AS CAST.

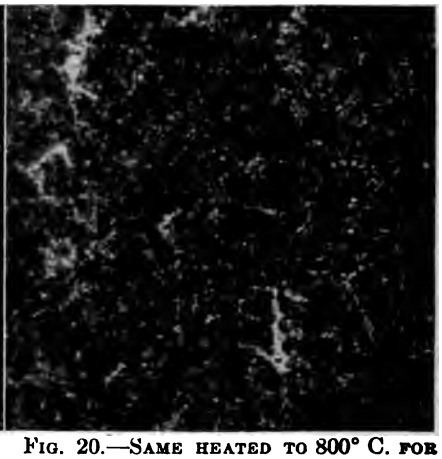

FIG. 20.—SAME HEATED TO 800° C. FOR 2 HR. AND COOLED IN LIME.

FIG. 21.—SAME HEATED TO 900° C. FOR 2 HR. AND COOLED IN LIME.

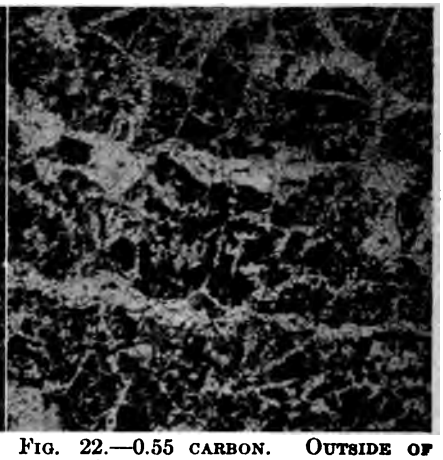

FIG. 22.—0.55 CARBON. OUTSIDE OF 6-IN. SQUARE BAR AS CAST.

FIG. 23.—SAME HEATED TO 800° C. FOR 2 HR. AND COOLED IN LIME.

FIG. 24.—SAME HEATED TO 900° C. FOR 2 HR. AND COOLED IN LIME.

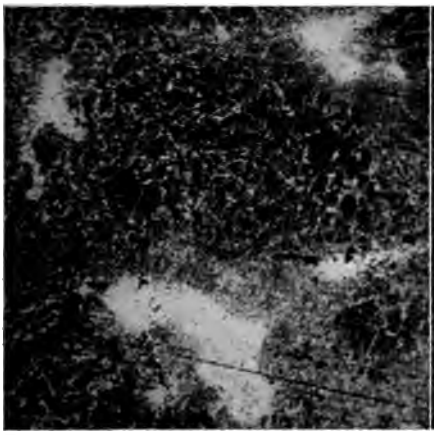

FIG. 25.—0.55 CARBON. CENTER OF 6-IN. SQUARE BAR AS CAST.

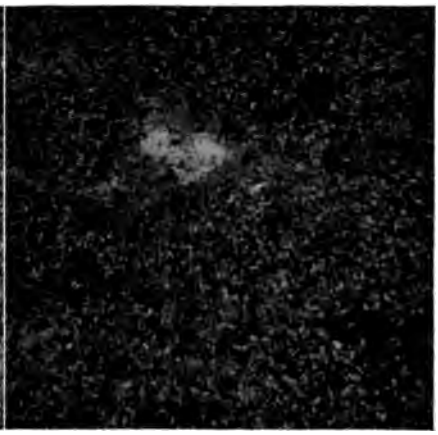

FIG. 26.—SAME HEATED TO 800° C. FOR 2 HR. AND COOLED IN LIME.

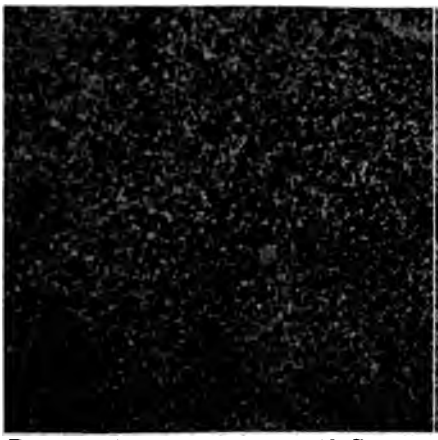

FIG. 27.—SAME HEATED TO 900° C. FOR 2 HR. AND COOLED IN LIME.

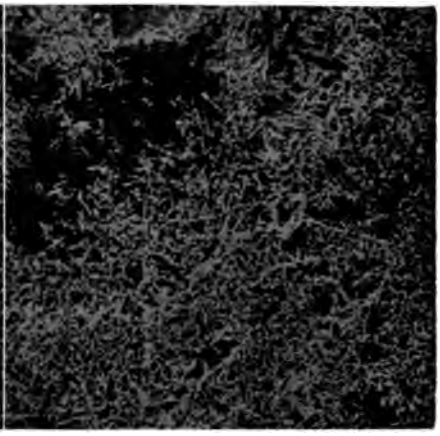

FIG. 28.—0.42 CARBON. ANNEALED AT 900° C. FOR 3 HR. AND COOLED IN AIR BLAST.

FIG. 29.—0.42 CARBON. ANNEALED AT 900° C. FOR 2 HR., COOLED SLOWLY TO 800° C. AND QUENCHED IN WATER.

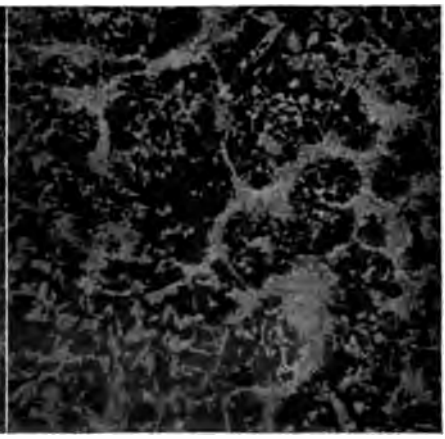

FIG. 30.—0.39 CARBON. ANNEALED AT 900° C. FOR 2 HR., COOLED SLOWLY TO 750° C. AND QUENCHED IN WATER.

As a result of this experiment, we selected 900° C. as the best annealing temperature for all of our castings, regardless of the carbon or the thickness of the casting; 800° C. was apparently not high enough to eliminate all of the free ferrite without extremely long annealing and both 900° C. and 950° C. were high enough to accomplish the result desired in 2 hr. without undue coarsening of the microstructure. Theoretically, 900° C. should be too high for the 0.55 per cent. carbon steel, but as our investigations have shown very little coarsening of the grain either at 900° C. or at 950° C., we have never thought it worth while to vary the annealing temperature for variations in carbon. The correctness of this conclusion has been borne out by 11 years of commercial practice.

When we had completed the work just described, we proceeded to anneal our castings by heating to 900° C. for 2 hr. or a little over and cooling them in the furnace. The steel so annealed showed a microstructure like that of Figs. 30 and 31, which it is apparent at a glance are entirely different from the appearance of the small pieces annealed at the same temperature and cooled in lime at the laboratory. It immediately became evident that the slower cooling allowed the separation of the ferrite, which had been suppressed in the small specimens by the comparatively rapid cooling in the lime box.

Effect of Rate of Cooling on Separation of Ferrite

In order to determine definitely the effect of the rate of cooling on the separation of ferrite, a number of bars were annealed at 900° C. for 2 hr. or slightly over and cooled at various rates. Fig. 28 shows the effect of cooling a bar rapidly in a blast of air and Fig. 34 shows the effect of cooling undisturbed in air. It will be noticed that in both of these the ferrite has separated into a very fine network, which resembles closely the structure of the small pieces annealed and cooled in the lime box. The only difference between these two is that the specimen cooled in the blast of air shows small areas of sorbite due to the more rapid cooling. Fig. 29 shows the structure of a bar annealed at 900° C., cooled slowly in the furnace to 800° C., and quenched in water; and Figs. 30 and 31 show the structures of two bars annealed at the same temperature and length of time and cooled slowly to 750° C. and 500° C., respectively, and then quenched. It will be seen at once that in cooling slowly from 800° C. to 750° C., the major part of the ferrite has separated out and that this ferrite tends strongly to occupy the position it occupied in the steel in the untreated condition. That is, there is a strong tendency for the ferrite to occupy the old grain boundaries and thus lead to structural weakness.

We believe that several investigators who have studied this question and carried on their work entirely in a laboratory have failed to take account of this fact that we have just brought out, that really slow cooling

allows the separation of free ferrite and that it is not possible to anneal cast steel and cool it slowly in the furnace and get a fine-grained microstructure unless, of course, the carbon is around 0.80 per cent. in which case there is no free ferrite to separate out. It has been stated in more than one paper that annealing followed by slow cooling gives cast steel a fine-grained microstructure unless there is something wrong with the steel, such as high phosphorus or sulfur; but our experience shows clearly that slow cooling after annealing always brings out free ferrite in large masses regardless of the phosphorus, the sulfur, or the process by which the steel is made, and there are in our files hundreds of micrographs and tests to show this fact.

Reheating Temperature for Castings Cooled in Air after Annealing

Having ascertained that in order to retain in the steel a fine microstructure it was necessary to cool the castings at least as rapidly as in air, we proceeded to determine the proper temperature to which the steel should be reheated in order to eliminate the cooling stresses and the hardening effect of the air cooling. We thought at first that in order to accomplish these two results it would be necessary to reheat the steel above Ac_1. Figs. 32 and 34 show the microstructures of a bar of steel annealed for 4 hr. at 900° C. and cooled in the air and afterward reheated to 760° C. at one end, and not over a black heat at the other, and cooled slowly in the furnace. It will be seen at once that where the steel attained a temperature of 760° C. considerable free ferrite separated out and that it separated out along the old grain boundaries, more or less in the positions it occupied in the untreated steel. We show only three specimens from this bar, of which we took in all eight micrographs. Two sections between Figs. 32 and 33 and two cuts between Figs. 33 and 34 showed a lessening amount of free ferrite so that we proved to ourselves definitely that as the temperature fell off below 760° C. the amount of free ferrite separating out was rapidly reduced. This fact has been proved beyond the shadow of a doubt by the examination of several hundred castings annealed in practice, although at the time this research was made we were in doubt for some time as to whether this free ferrite was due to reheating to too high a temperature or to insufficient heating on the first anneal.

Figs. 35 and 36 show the structure of two bars of steel annealed at 900° C. for 4 hr. and cooled in the air and then reheated to 700° C. for 30 min. and 4 hr., respectively, and cooled slowly in the furnace. It will be observed that the fine network structure has been retained perfectly. For purposes of comparison we have added Fig. 37, which shows the

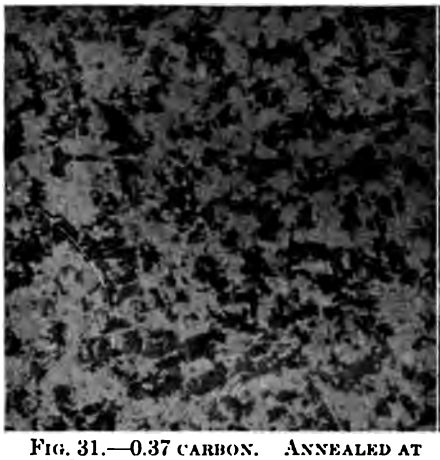

FIG. 31.—0.37 CARBON. ANNEALED AT 900° C. FOR 2 HR., COOLED SLOWLY TO 500° C. AND QUENCHED IN WATER.

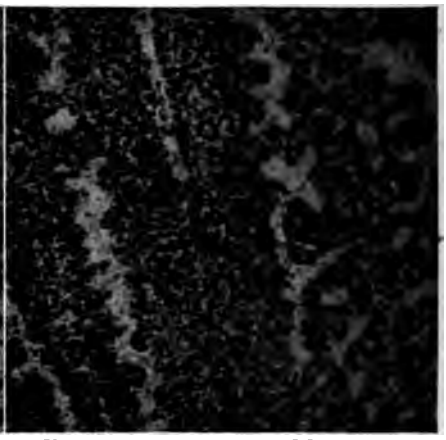

FIG. 32.—0.50 CARBON. METCALF BAR ANNEALED AT 900° C. FOR 4 HR. AND AIR COOLED. REHEATED TO 760° C. AND SLOWLY COOLED IN FURNACE.

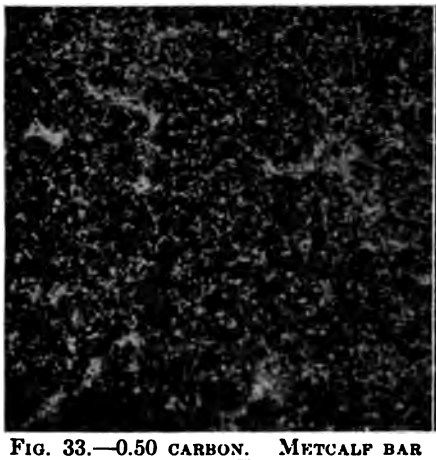

FIG. 33.—0.50 CARBON. METCALF BAR 4 IN. FROM FIG. 32.

FIG. 34.—0.50 CARBON. METCALF BAR 6 IN. FROM FIG. 32.

FIG. 35.—0.50 CARBON. ANNEALED AT *900° C. FOR 4 HR.* AND AIR COOLED. *REHEATED TO 700° C.* FOR 30 MIN. AND *SLOWLY COOLED IN FURNACE.*

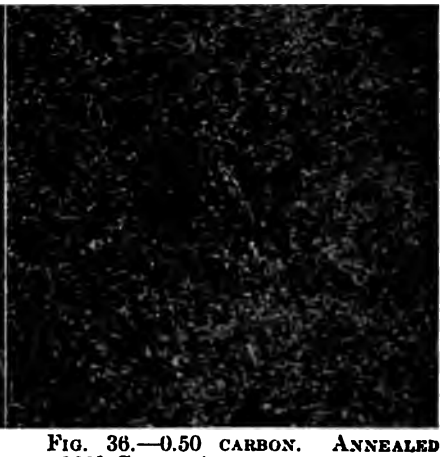

FIG. 36.—0.50 CARBON. ANNEALED AT 900° C. FOR 4 HR. AND AIR COOLED. REHEATED TO 700° C. FOR 4 HR. AND SLOWLY COOLED IN FURNACE.

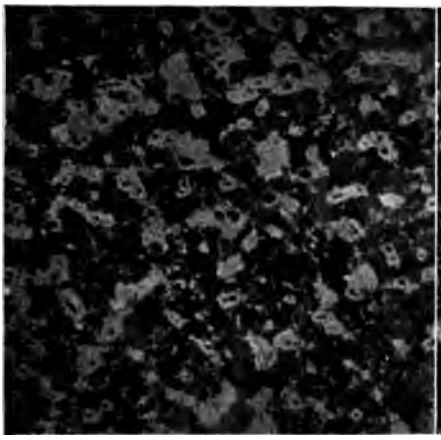

FIG. 37.—0.50 CARBON. ANNEALED AT 900° C. FOR 4 HR. AND SLOWLY COOLED IN FURNACE.

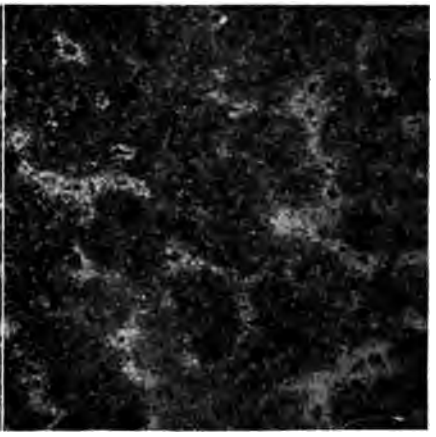

FIG. 38.—0.50 CARBON. ANNEALED AT 900° C. FOR 4 HR. AND AIR COOLED. REHEATED TO 760° C. FOR 5 HR. AND SLOWLY COOLED IN FURNACE.

FIG. 39.—0.59 CARBON. ANNEALED AT 900° C. FOR 4 HR. AND AIR COOLED. REHEATED TO 760° C. FOR 2 HR. AND SLOWLY COOLED IN FURNACE. P, 0.017; S, 0.023.

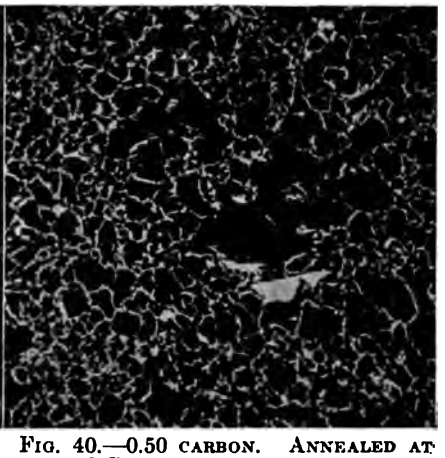

FIG. 40.—0.50 CARBON. ANNEALED AT 1000° C. FOR 7 HR. AND AIR COOLED.

FIG. 41.—0.50 CARBON. ANNEALED AT 1000° C. FOR 2 HR. AND AIR COOLED. *REHEATED TO 760° C. FOR 4 HR. AND SLOWLY COOLED.*

FIG. 42.—0.50 CARBON. ANNEALED AT 1000° C. FOR 7 HR. AND AIR COOLED. REHEATED TO 760° C. FOR 4 HR. AND SLOWLY COOLED.

structure of steel cooled slowly in the furnace after annealing at 900° C., and Fig. 38, which shows steel cooled in air after annealing at 900° C. and then reheated to 760° C. for 5 hr. and cooled slowly. Fig. 39, even though of higher carbon, shows exactly the same thing as Fig. 38 and

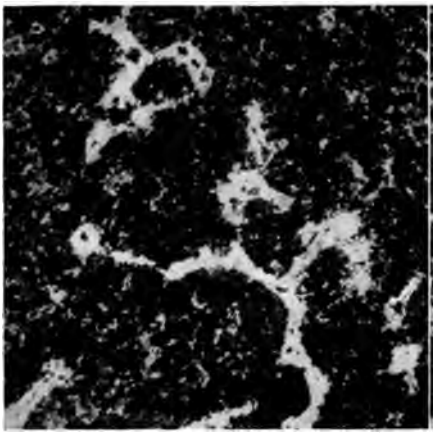

FIG. 43.—0.50 CARBON. 3½-IN. SQUARE UNFORGED BESSEMER STEEL. P, 0.051. ANNEALED AT 900° C. FOR 5½ HR. AND AIR COOLED. REHEATED TO 760° C. FOR 7½ HR. AND SLOWLY COOLED IN FURNACE.

FIG. 44.—0.50 CARBON. SAME BESSEMER STEEL FORGED TO 2-IN. SQUARE. SAME TREATMENT.

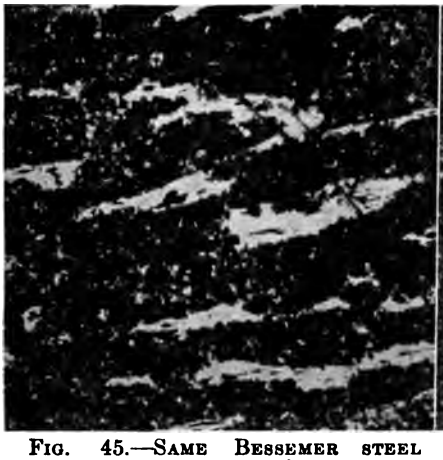

FIG. 45.—SAME BESSEMER STEEL FORGED TO 1-IN. SQUARE. SAME TREATMENT.

FIG. 46.—0.54 CARBON. 4-IN. SQUARE UNFORGED ELECTRIC STEEL. P, 0.028. ANNEALED AT 900° C. FOR 2 HR. AND AIR COOLED. REHEATED TO 760° C. FOR 6 HR. AND SLOWLY COOLED IN FURNACE.

will be referred to later in discussing the effect of lower phosphorus and sulfur on the separation of free ferrite. In Table 1 are given the averages of many tests from castings whose treatments and microstructures correspond to Figs. 36, 37, and 38. These tests will be discussed later.

Reheating Temperature for Castings Quenched in Water or Oil after Annealing

During a period of several months we manufactured steel castings that were heat treated to give a microstructure similar to Figs. 35 and 36

Fig. 47.—0.54 carbon. Same electric steel forged to 2⅛-in. square. Same treatment as Fig. 46.

Fig. 48.—Same electric steel forged to ½-in. square. Same treatment as Fig. 46.

Fig. 49.—0.54 carbon. 4-in. square unforged electric steel. P, 0.028. Annealed at 900° C. for 2 hr. and quenched in water. Reheated to 760° C. for 2 hr. and slowly cooled in furnace.

Fig. 50.—0.54 carbon. Same electric steel forged to 2⅛-in. square. Same treatment as Fig. 49.

and which gave tests of which line 3 in Table 1 shows average physical properties. It is perhaps just as well to state here that on one very important class of castings we examined and tested an attached coupon from each and every casting.

It was clear to us at this time (the fall of 1908) that we would secure even better microstructures and physical properties than those just illustrated by quenching the castings either in oil or in water and reheating them to a suitable temperature. We accordingly annealed several

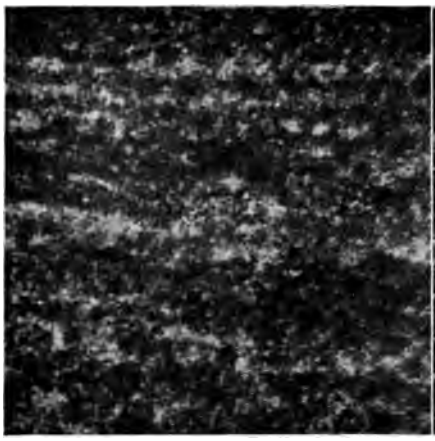

FIG 51.—SAME ELECTRIC STEEL FORGED TO ½-IN. SQUARE. SAME TREATMENT AS FIG. 49.

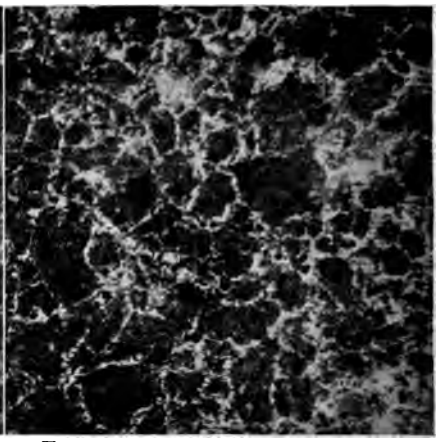

FIG. 52.—0.54 CARBON. 4-IN. SQUARE UNFORGED ELECTRIC STEEL. P, 0.028. ANNEALED AT 900° C. FOR 2 HR. AND COOLED IN AIR. REHEATED TO 680° C. FOR 6 HR. AND SLOWLY COOLED IN FURNACE.

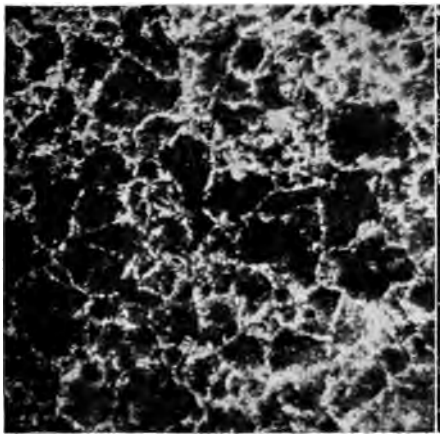

FIG. 53.—0.54 CARBON. SAME ELECTRIC STEEL FORGED TO 2⅛-IN. SQUARE. SAME TREATMENT AS FIG. 52.

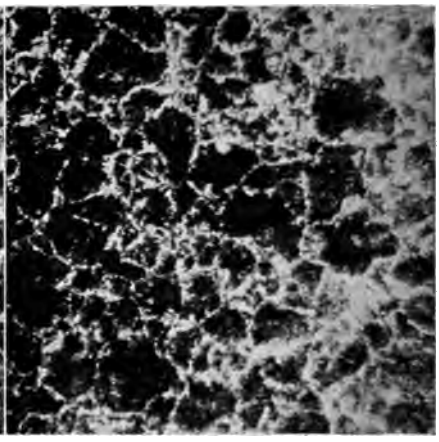

FIG. 54.—SAME ELECTRIC STEEL FORGED TO ½-IN. SQUARE. SAME TREATMENT AS FIG. 52.

bars at 900° C., which were then cooled slowly to 800° C. and quenched in water. Fig. 58 shows the microstructure of a bar so annealed and then reheated to 650° C. and again quenched; and Fig. 59 shows the microstructure of a bar annealed and quenched in the same way, and then

reheated to 700° C. and quenched in water. It is at once apparent that upon reheating to 700° C. there is a strong tendency for the ferrite to separate out, just as it did in the air-cooled steel on reheating to 760° C. At the time these experiments were made, there was considerable dis-

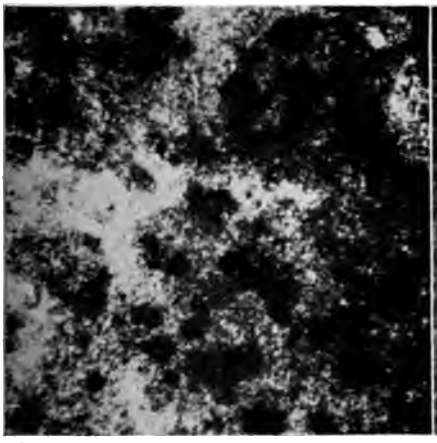

FIG. 55.—0.54 CARBON. 4-IN. SQUARE UNFORGED ELECTRIC STEEL. P, 0.028 ANNEALED AT 900° C. FOR 2 HR. AND QUENCHED IN WATER. REHEATED AT 660° C. FOR 6 HR. AND SLOWLY COOLED IN FURNACE.

FIG. 56.—0.54 CARBON. SAME ELECTRIC STEEL FORGED TO 2⅛-IN. SQUARE. SAME TREATMENT AS FIG. 55.

FIG. 57.—0.54 CARBON. SAME ELECTRIC STEEL FORGED TO ½-IN. SQUARE. SAME TREATMENT AS FIG. 55.

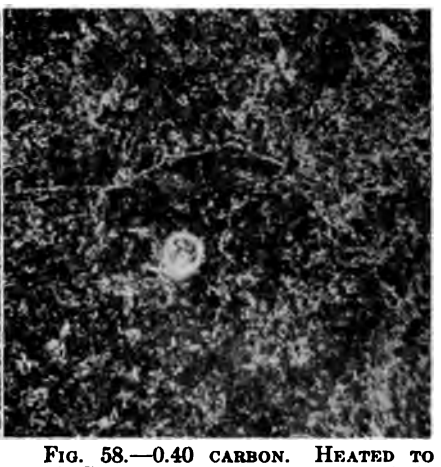

FIG. 58.—0.40 CARBON. HEATED TO 900° C., COOLED SLOWLY TO 800° C. AND QUENCHED IN WATER. REHEATED TO 650° C. AND QUENCHED.

cussion as to whether this reappearance of free ferrite at 700° C. was characteristic or whether it was due to insufficient annealing at 900° C.

We found after considerable experimenting that the physical properties and microstructures of the test bars were practically the same for

bars quenched in water as for bars quenched in oil and that it was desirable to quench in oil only when we were afraid that the more drastic water quenching would crack the castings; in this paper we make no attempt to discriminate between the microstructures of the oil-quenched and the water-quenched test bars.

FIG. 59.—0.40 CARBON. HEATED TO 900° C., SLOWLY COOLED TO 800° C. AND QUENCHED IN WATER. REHEATED TO 700° C. AND QUENCHED IN WATER.

FIG. 60.—0.50 CARBON. METCALF BAR ANNEALED AT 900° C. FOR 4 HR. AND QUENCHED IN OIL. THIS SECTION IS 1½ IN. FROM HOT END WHICH WAS REHEATED TO 730° C. AND SLOWLY COOLED IN FURNACE.

FIG. 61,—0.50 CARBON. SAME METCALF BAR. SECTION 4¾ IN. FROM HOT END.

FIG. 62.—0.50 CARBON. SAME METCALF BAR. SECTION 5½ IN. FROM HOT END.

In order to demonstrate that the reappearance of free ferrite on reheating the quenched steel to 700° C., or above, is characteristic, we resorted again to the ever useful method of Metcalf and Figs. 60–62 are selected from seven micrographs made from a bar annealed at 900° C. for

4 hr. and quenched in oil and then reheated to 730° C. at one end and kept as cold as possible at the other, and then cooled slowly in the furnace. The appearance of free ferrite against a fine-grained background in the specimen heated to 730° C. is apparent at a glance in Fig. 60 while Fig. 61, taken 4¾ in. (12 cm.) from the hot end, shows virtually no free ferrite, and Fig. 62, 5½ in. (14 cm.) from the hot end, shows none. We then, of course, proceeded to find the proper reheating temperature. Figs. 63 and 64 show the structures of oil-quenched bars reheated to 660° C. for 2 hr. and to 680° C. for 6 hr., respectively, and then cooled slowly in the furnace. It is quite clear that the separation of free ferrite in these specimens has been almost, if not entirely, prevented. These results have been confirmed by the examination of a considerable number of castings heat treated on a commercial scale. Line 6 of Table 1 shows the physical properties characteristic of this grade of steel heat treated in this manner.

Comparison of Drawing Temperature Versus Quenching Medium

It seemed to eminent authorities so inherently improbable that free ferrite would separate from the oil-quenched steel at a lower temperature than from the air-cooled steel that we were requested to prove this point even more definitely and we did so on two sets of test bars. Figs. 65 and 66 show the microstructures of two bars of the same heat of steel annealed side by side in the same annealing furnace at the laboratory to 900° C. for 3 hr. One was then cooled to 850° C. and quenched in oil, the other was air cooled from 900° C. Both were heated, side by side, to 720° C. for 2 hr. and one was quenched in oil and the other cooled slowly in the furnace. It needs but a glance to show that the free ferrite has separated strongly in the oil-quenched bar while there is no separation at all in the bar cooled in the air. For comparison with these two specimens, Fig. 59, which has already been described, shows the separation of free ferrite on reheating a water-quenched test bar to 700° C. and again quenching.

Influence of Other Treatments

The work so far described was completed early in 1909. At about that time our attention was called to some work that had been done by other investigators, who used a temperature for annealing as high as 1000° C., and we were urged to assure ourselves that the reappearance of free ferrite in our experiments was really characteristic and not due to insufficient annealing before the rapid cooling. We accordingly annealed a number of bars at 1000° C. for varying lengths of time. Fig. 40 shows the microstructure of one of these bars annealed for 7 hr. at 1000° C. and cooled in air. The microstructure of this bar is practically the same as that of steel annealed at 900° C. and cooled in air, Fig. 34, ex-

cept that the grains are considerably coarser, due to the higher annealing temperature. Out of the sixteen test bars annealed for various lengths of time and reheated for various lengths of time, we have selected Figs. 41 and 42 as characteristic. Fig. 41 shows the microstructure of a bar an-

FIG. 63.—0.50 CARBON. ANNEALED AT 900° C. FOR 4 HR. AND QUENCHED IN OIL. REHEATED TO 660° C. FOR 2 HR. AND SLOWLY COOLED IN FURNACE.

FIG. 64.—0.50 CARBON. ANNEALED AT 900° C. FOR 4 HR. AND QUENCHED IN OIL. REHEATED TO 680° C. FOR 6 HR. AND SLOWLY COOLED IN FURNACE.

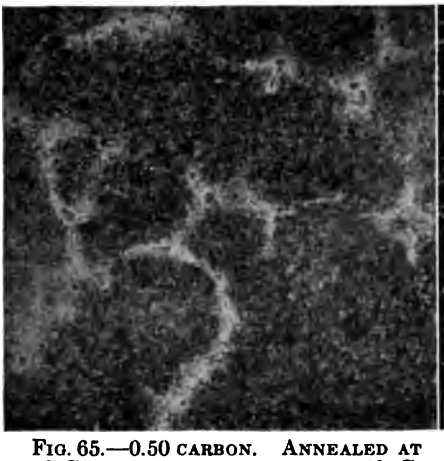

FIG. 65.—0.50 CARBON. ANNEALED AT 900° C. FOR 3 HR. COOLED TO 850° C. AND QUENCHED IN OIL. REHEATED TO 720° C. FOR 2 HR. AND QUENCHED IN OIL.

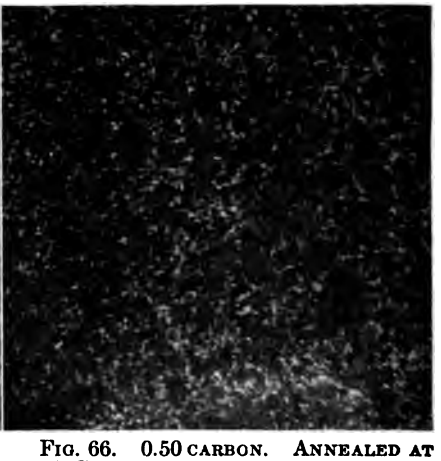

FIG. 66. 0.50 CARBON. ANNEALED AT 900° C. FOR 3 HR. AND AIR COOLED. REHEATED TO 720° C. FOR 2 HR. AND SLOWLY COOLED.

nealed at 1000° C. for 2 hr., cooled in air, reheated to 760° C. for 4 hr. and cooled slowly; and Fig. 42 shows the microstructure of a similar bar annealed at 1000° C. for 7 hr. and cooled in air, reheated to 760° C. for 4 hr., and cooled slowly. The reappearance of the free ferrite is too plain to leave any doubt in our minds that the annealing at 1000° C.

has had no more effect in suppressing ferrite than has annealing at 900° C. We annealed a number of castings on a commercial scale at 1000° C., the examination of which absolutely confirmed this conclusion. These results indicated clearly that there was no possible need of using the higher annealing temperature, especially in view of the fact that annealing at 1000° C. would use much more fuel, scale the castings much more heavily, and very much increase the cost of up-keep of the annealing furnaces.

Although we had no hope of using such a procedure in practice, we thought it would be interesting to try quenching a piece of cast steel at or above 900° C. while it retained the initial heat of casting. Our idea was to ascertain whether the result of preventing the actual formation of primary ferrite in this bar on cooling to room temperature would aid in suppressing the reappearance of free ferrite on reheating the steel above Ac_1. Figs. 67 and 68 show the microstructures of the outside and center of the bar of steel quenched in water at about 950° C. while cooling from the molten state. Apparently Fig. 67 shows martensite and sorbite and Fig. 68 shows the same constituents and a little free ferrite, although we do not feel absolutely certain of the identification of these constituents.

Fig. 69 shows the microstructure of the same bar of steel after reheating to 760° C. for 2 hr. and cooling slowly in the furnace. It is at once apparent that the free ferrite has separated out in the same characteristic manner as in our previous experiments, and we concluded that there was no use in going further along these lines, although from a scientific standpoint it would undoubtedly have been interesting to make more extensive experiments.

Effect of Forging

Although it was not quite apparent how it could be applied in practice, we desired to ascertain what would be the effect of a certain amount of forging or pressing of the important sections of a steel casting before heat treatment. We accordingly forged a 3½-in. (8.9-cm.) square cast billet, with a composition of carbon 0.49 per cent., silicon 0.39 per cent., manganese 0.76 per cent., sulfur 0.055 per cent., and phosphorus 0.051 per cent., under the steam hammer in such a way that we had, at one point, the original cross-section and at the other points cross-sections of 3 in., 2¼ in., 2 in., 1¾ in., 1½ in., and 1 in., respectively. The entire piece of steel was then annealed at 900° C. for 5½ hr., cooled in air, reheated to 760° C. for 7½ hr., and cooled slowly in the furnace. Longitudinal sections were made from each step in the forging, of which we show only three. Fig. 43 shows the original cast steel, Fig. 44 shows the 2-in. (5-cm.) square section, and Fig. 45 shows the 1-in. square section. It is apparent at once that the free ferrite reappears not only in the cast

steel but in the forged steel, even in the 1-in. square section; and that the only difference in the appearance of the ferrite is that the forging has drawn it out into long streaks closely resembling the well-known ferrite ghosts, which have been so extensively described in metallurgical

FIG. 67.—0.48 CARBON. OUTSIDE OF BAR QUENCHED IN WATER AT ABOUT 950° C. WHILE COOLING FROM MOLTEN STATE. MARTENSITE + SORBITE.

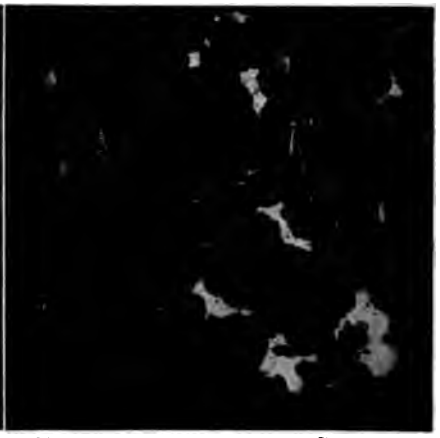

FIG. 68.—0.48 CARBON. CENTER OF SAME BAR AS FIG. 67. MARTENSITE + SORBITE + FERRITE.

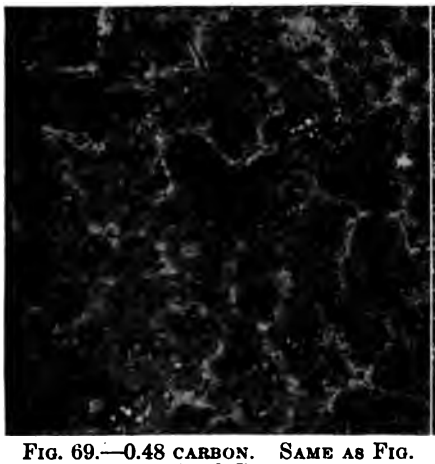

FIG. 69.—0.48 CARBON. SAME AS FIG. 68 REHEATED TO 760° C. FOR 2 HR. AND SLOWLY COOLED IN FURNACE.

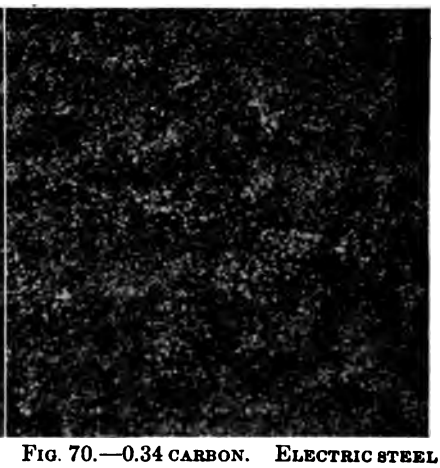

FIG. 70.—0.34 CARBON. ELECTRIC STEEL AS RECEIVED.

literature. These experiments were made late in 1908 and, as will readily be inferred, they completely disposed of the idea that we could improve Bessemer cast steel by light pressing or forging before heat treatment.

When we selected the test data for this paper, it seemed worth while

to carry this experiment a little further. We accordingly prepared a 4-in. square cast billet of basic electric steel of the following analysis: carbon 0.54 per cent.; silicon 0.44 per cent.; manganese 0.88 per cent.; sulfur 0.028 per cent., and phosphorus 0.028 per cent. This billet was forged[1] in a similar manner to the Bessemer billet to sections of 3 in. (7.6 cm.), 2½ in., 2⅛ in., 1⅝ in., 1¼ in., ¾ in., and ½ in. square, respectively. Our idea was to determine, if possible, whether a lower phosphorus and sulfur and a different method of manufacture would have much effect on the characteristic behavior of the steel in heat treatment.

The electric steel treated in the same manner as the Bessemer steel is shown in Figs. 46, 47 and 48. Fig. 46 shows the cast steel, Fig. 47 shows the steel reduced to 2⅛ in. square, and Fig. 48 shows the steel forged down to ½ in. square. It will be seen that the separation of the ferrite has taken place, but apparently not to as marked an extent as in the case of the Bessemer steel. It should be stated, however, that visual examination of these specimens shows the streaked structure much more clearly than is indicated in these micrographs. Figs. 49, 50, and 51 show this same electric steel reheated to 760° C. for 2 hr. after quenching in water, following an anneal of 2 hr. at 900° C. Fig. 49 shows the cast steel, Fig. 50 the 2⅛-in. square, and Fig. 51 the ½-in. square. The separation of free ferrite is again plainly shown and, as in the case of the previous three specimens, visual examination shows this much more clearly than the micrographs.

In order to show that the ferrite can be largely suppressed by proper drawing temperatures, we treated the same pieces as shown above, using lower temperatures for the second heating. Figs. 52, 53, and 54 show the cast steel, the 2⅛-in. (5.38-cm.) square and the ½-in. (1.27-cm.) square, respectively, after annealing at 900° C. for 2 hr., cooling in air, reheating to 680° C. for 6 hr. and cooling slowly in the furnace. It will be noticed that the separation of free ferrite has been almost, if not entirely, suppressed. Figs. 55, 56, and 57 show the same sections after annealing for 2 hr., at 900° C., quenching in water, reheating at 660° C. for 6 hr., and cooling slowly in the furnace. The separation of free ferrite has been practically suppressed, although Fig. 55 would indicate a considerable separation. Visual examination of the specimen shown in Fig. 55 indicated a much more uniform microstructure than this particular micrograph, as the ferrite spots shown in that figure are the only large ones found in the entire cross-section. Fig. 57, and especially Fig. 56, show a slight separation of free ferrite. In this connection our experience has shown that it is not always possible entirely to prevent the separation of a certain amount of free ferrite on reheating cast steel that has been quenched in either oil or water.

[1] Forging temperature 1150° C.

Some further light is shed on this question of the connection between ferrite ghosts and low phosphorus and sulfur by Fig. 39 and work we have done on steel represented by Fig. 81. Fig. 39 shows a bar of cast steel with 0.017 per cent. phosphorus and 0.023 per cent. sulfur, which was annealed for 4 hr. at 900° C., cooled in air, reheated to 760° C. for 2 hr., and cooled slowly in the furnace. There is unfortunately no record of the origin of this specimen, but it was procured at the time we did most of this work with a view to ascertaining whether lower phosphorus and sulfur would result in the suppression of free ferrite on heat treating. The micrograph shows that in this steel, also, free ferrite separates out, if the second heating is carried above Ac_1.

Figs. 75 and 81 show a cast coupon of electric steel properly heat treated in which very little free ferrite has separated. Recently we hunted up this piece of steel in our specimen cases, and also several other specimens of the same series, heated them to 760° C. for 6 hr. and cooled them slowly in the furnace. The analyses of these several pieces of steel are given in lines 10, 4, and 7 of Table 4. Attention is particularly directed to the extremely low figure for phosphorus and sulfur content of these specimens. We found on reheating these specimens to 760° C., as just described, that the free ferrite separates out; but a careful visual examination of seven specimens indicates that the separation is not nearly so pronounced as it is in the Bessemer steels described in this paper, or in the steel of Fig. 39. The indications are, therefore, that in the case of very low-phosphorus and sulfur steels made in the electric furnace, the separation of free ferrite on reheating above Ac_1 takes place to a smaller extent than in the case of higher phosphorus and sulfur material.

In this connection, however, we would call attention to the fact that these steels, after annealing at about 900° C. and cooling slowly, show a coarse ferrite structure, practically the same as the structure of the Bessemer steels already discussed. Thus Figs. 73, 76, and 79 show the microstructures of cast electric steel sent to us from outside makers and quite evidently annealed in the usual manner by heating to about 900° C. and cooling slowly in the furnace. The coarse microstructure is clearly indicated. Fig. 70 shows the microstructure of another sample of electric steel made by outside parties, as received by us. We believe both the microstructure and the tests demonstrate that this particular steel was given a double annealing by the maker. The improvement in the physical properties of these electric steels by double treatment will be discussed a little later, when it will be shown quite clearly that the physical properties of the steel in the fine-grained condition are considerably better than those of the steel with a coarse microstructure.

We do not wish to be understood as stating categorically that the phosphorus and sulfur contents of the steel have nothing to do with the appearance of ferrite ghosts, but we would call attention to the fact that

if ghosts occur strongly in steel containing but 0.028 per cent. phosphorus and sulfur, it is important that the heat-treatment temperatures be properly controlled, because the greater part of the steel castings produced in this country, even from the electric furnace, contain at least 0.03 per

FIG. 71.—0.34 CARBON. SAME ELECTRIC STEEL ANNEALED AT 900° C. FOR 3 HR. AND QUENCHED IN WATER. REHEATED TO 760° C. AND COOLED IN AIR.

FIG. 72.—0.34 CARBON. SAME ELECTRIC STEEL ANNEALED AT 900° C. FOR 3 HR. AND COOLED IN AIR. REHEATED TO 720° C. FOR 6 HR. AND COOLED IN AIR.

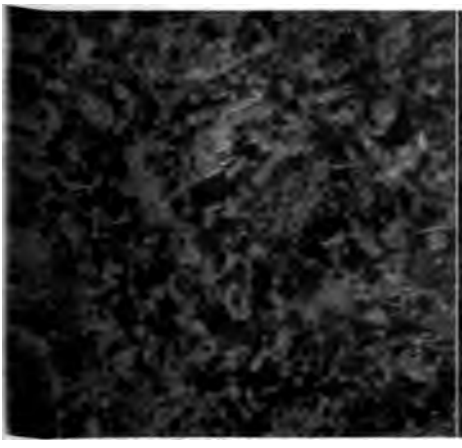

FIG. 73.—0.37 CARBON. ELECTRIC STEEL AS RECEIVED.

FIG. 74.—0.37 CARBON. SAME ELECTRIC STEEL ANNEALED AT 900° C. FOR 3 HR. AND QUENCHED IN WATER. REHEATED TO 650° C. FOR 6 HR. AND COOLED IN AIR.

cent. phosphorus and sulfur, and generally considerably more. It is also self-evident that it is considerably cheaper to heat-treat the steel in such a way as to suppress the ghosts than it would be to try to purify steel below 0.02 per cent. in phosphorus and sulfur. While we do not wish to state this unequivocally, our tests running over a period of 13

years have indicated that if the ghosts are suppressed they have little or no effect on the physical properties of the steel; whereas if the steel is so heat treated as to produce ghosts (single treatment or too high drawing temperatures) the bad effect on the physical properties is very marked.

FIG. 75.—0.37 CARBON. SAME ELECTRIC STEEL ANNEALED AT 900° C. FOR 3 HR. AND COOLED IN AIR. REHEATED TO 720° C. FOR 6 HR. AND COOLED IN AIR.

FIG. 76.—0.39 CARBON. ELECTRIC STEEL AS RECEIVED.

FIG. 77.—0.39 CARBON. SAME ELECTRIC STEEL ANNEALED AT 900° C. FOR 3 HR. AND QUENCHED IN WATER. REHEATED TO 660° C. FOR 8 HR. AND COOLED IN AIR.

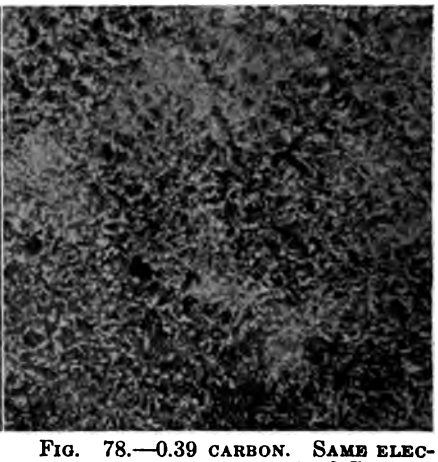

FIG. 78.—0.39 CARBON. SAME ELECTRIC STEEL ANNEALED AT 900° C. FOR 3 HR. AND COOLED IN AIR. REHEATED TO 710° C. FOR 6 HR. AND COOLED IN AIR.

COMPARISON OF PHYSICAL PROPERTIES VERSUS TREATMENTS

In Table 1, lines 1, 2, and 3 show the averages of a great number of tests covering approximately 0.50 per cent. carbon steel heat treated in three ways. Line 1 gives the tests obtainable by heating to 900° C. for

TABLE 1.—*Bessemer Steel*

Line	Carbon, Per Cent.	Silicon, Per Cent.	Manganese, Per Cent.	Tensile Strength, Lb. per Sq. In.	Elastic Limit, Lb. per Sq. In.	Elongation, Per Cent.	Contraction, Per Cent.	Fremont	No. of Tests	Treatment
1	0.50	0.54	0.59	83,500	43,600	19.8	24.5	5.5[a]	11	900° C (4) furnace cooled
2	0.51	0.38	0.56	83,200	41,650	19.5	19.2	5.8	21	900° C (4) air 760° C (6) furnace
3	0.51	0.44	0.69	83,650	47,100	22.5	26.6	5.9	79	920° C (5) air 700° C (6) furnace
4	0.42	0.43	0.58	76,500	39,700	22.0	25.0	6.5[b]	157	900° C (4) furnace cooled
5	0.39	0.44	0.73	79,100	41,200	26.2	35.7	8.8	32	900° C (4½) air 700° C (6) furnace
6	0.42	0.54	0.71	81,450	49,590	23.9	37.9	20.5	2	900° C (4) oil 680° C (6) furnace
7	0.42	0.54	0.71	81,700	50,720	26.4	44.2	17.7	2	900° C (4) water 680° C (6) furnace
8	0.32	0.44	0.65	68,550	40,500	31.8	50.7	17.4	9	900° C (4½) air 700° C (8) furnace
9	0.31	0.52	1.12	81,190	52,380	29.5	56.9	28.4	5	900° C (4) water 700° C (6) furnace
10	0.27	0.46	0.71	72,140	45,930	32.9	57.6	35.5	4	900° C (4) water 700° C (6) furnace
11	0.26	0.45	1.11	78,100	50,700	30.5	56.4	25.6	29	900° C (4) water 700° C (6) furnace
12	0.26	0.51	1.06	82,240	51,360	29.0	55.9	25.6	7	880° C (5) water 615° C (8) furnace
13	0.26	0.56	1.17	83,250	51,200	29.8	58.2	27.6	21	900° C (5) water 630° C (8½) furnace
14	0.21	0.43	1.07	75,500	49,500	30.7	56.2	27.1	21	900° C (4) water 700° C (6) furnace
15	0.22	0.53	1.08	81,100	54,450	29.8	57.9	27.6	12	900° C (4) water 650° C (6) furnace
16	0.22	0.51	1.08	81,520	53,300	29.5	57.3	26.3	52	900° C (6) water 620° C (8) furnace

[a] Average of 9 tests.

[b] Average of 22 tests.

TABLE 2.—*Acid Open-hearth Steel*

Line	Carbon, Per Cent.	Silicon, Per Cent.	Manganese, Per Cent.	Tensile Strength, Lb. per Sq. In.	Elastic Limit, Lb. per Sq. In.	Elongation, Per Cent.	Contraction, Per Cent.	Fremont	No. of Tests	Treatment
1	0.40	0.33	0.72	82,800	44,800	22.6	32.5	....	12	900° C. furnace cool.
2	0.34	0.31	0.70	83,000	43,700	22.5	37.5	....	6	900° C. furnace cool.
3	0.24	0.30	0.63	69,600	35,800	30.0	45.2	....	14	900° C. furnace cool.

4 hr. and cooling slowly in the furnace, corresponding to the microstructures indicated by Figs. 31 and 37. Line 2 gives the tests obtainable by heating to 900° C. for 4 hr. and cooling in air, reheating to 760° C. for 6 hr. and cooling in the furnace. This treatment brings out free ferrite and is represented by Figs. 32, 38, and 39 and others. Line 3 gives the tests obtainable by heating to about 900° C. for 5 hr. and cooling in air, reheating to 700° C. for 6 hr. and cooling in the furnace and illustrated by the microstructure in Figs. 35, 36, and others. Of the three treatments indicated, the tests of the last give the best all-around properties. In nearly all of the tables we have included the Fremont shock test, for which we do not claim any great accuracy or uniformity in results. However, we have always used this test and are including it for what it may be worth.

The next comparison that can be made for similar carbons is represented by lines 4, 5, 6, and 7. This gives four different treatments for approximately 0.40 per cent. carbon steel treated as indicated in the table. The double treatment shown in lines 5, 6, and 7 gives a better elongation and very much better contraction of area and shock toughness than the plain single annealing. The quenched specimens show characteristically greater elastic limits as well as increased ductility. The balance of the table shows no comparison between the ordinary single treatment or plain anneal versus double treatment by either air or water quenching, but is included to show test results that can be very uniformly obtained for the compositions given by the application of double treatment and which we believe are very much better than any obtainable by the use of the plain anneal.

Tables 2 and 3 are averages of tests made by several other companies and are representative of the plain anneal with a probable hastened cooling in some of the cases. These tests are included so that comparisons may be made between steels of similar analyses, but made by different processes and heat treated in different manners as indicated. A general comparison will show that Bessemer steel double treated gives uniformly better all-around tests than plain annealed acid open-hearth or side-blown converter steel.

It is sometimes asserted that extremely soft steels require no heat treatment at all. We have never engaged extensively in the manufacture of this class but have made a limited number of tests on a Bessemer steel of the following analysis: carbon 0.10 per cent., silicon 0.19 per cent., manganese 0.23 per cent. These tests indicate that the tensile strength is increased by annealing and that the elastic limit is increased to a considerably greater extent. The contraction of area and the bend are also improved. The Fremont tests show the greatest improvement, running from 2.5 in the cast steel, 10.5 and 20.0 in the steel annealed and slowly cooled, to 17.5 in the air-cooled and drawn and 25.0 and 27.5 in the water-quenched. It is interesting to note in the last two tests

that the one that gave 27.5 was not reheated after quenching. Our tests on this steel are not of sufficient extent to justify the drawing of hard and fast conclusions but are interesting as indicating the beneficial results of double treatment.

In this connection we have frequently noticed descriptions of the practice in some shops where the steel is annealed at about 900° C. and slowly cooled in the furnace and afterward reheated to a lower temperature, generally at or below Ac_1. It is somewhat difficult to understand the reasoning on which such practice is based, because no structural change can take place in the steel by reheating below Ac_1 and little if any change could be expected from reheating somewhat above Ac_1. Consequently, one would expect the physical properties to be practically unaffected by the second heating unless the first heating was not properly carried out, so that some of the castings were improperly annealed in the first place. In that case, of course, some benefit might conceivably follow a second annealing, but the real remedy would be to carry out the first annealing properly. Neither could the second heating be of any use in relieving stresses, for the simple reason that the first annealing followed by slow cooling should completely eliminate the stresses in the castings.

Advantage of Double Treating Cast Steel

Table 4 gives the physical test results of the double treating of different electric steels made by other firms, some American and some European. Micrographs of these steels are shown in Figs. 70–81, inclusive. Lines 1, 2, and 3 and Figs. 70, 71, and 72 show the tests and structures obtained by treating in the manner indicated. It will be seen in this instance that the steel as received had in all probability already had a double treatment, and very little all-around improvement could be made in the properties of this steel by water or air quenching. The main improvement was in the bend, in which we secured 180° C. against 140° C., as received. The second steel, represented by lines 4, 5, and 6 and Figs. 73, 74, and 75, was greatly improved by the double treatment. The strength and elastic limit were on the average increased, the elongation remained nearly the same but the contraction of area, shock, and bend were greatly increased. The third steel, represented by lines 7, 8, and 9 and Figs. 76, 77, and 78, shows a great increase in strength and elastic limit, greater elongation and contraction, and very much greater shock toughness and bend. The fourth steel, represented by lines 10, 11, and 12 and Figs. 79, 80, and 81, shows the greatest increase in elastic limit, while the strength is also increased. The average elongation was improved slightly, the average contraction considerably, the shock toughness a little, and the bend considerably. In two cases, with water quenching, the bending test bar was bent flat on itself without breaking.

TABLE 3.—*Side-blown Converter Steel*

Line	Carbon, Per Cent.	Silicon, Per Cent.	Manganese, Per Cent.	Tensile Strength, Lb. per Sq. In.	Elastic Limit, Lb. per Sq. In.	Elongation, Per Cent.	Contraction, Per Cent.	Fremont	No. of Tests	Treatment
1	0.31	0.31	0.94	78,200	47,700	26.2	41.2	....	16	900° C. furnace cool.
2	0.27	0.28	0.80	72,500	42,500	29.1	48.1	....	57	900° C. furnace cool.
3	0.24	0.28	0.78	70,650	42,300	28.6	47.8	....	9	900° C. furnace cool.
4	0.20	0.40	0.73	68,930	43,140	27.2	39.7	....	10	900° C. furnace cool.

TABLE 4

Line	Carbon, Per Cent.	Silicon, Per Cent.	Manganese, Per Cent.	Sulfur, Per Cent.	Phosphorus, Per Cent.	Tensile Strength, Lb. per Sq. In.	Elastic Limit, Lb. per Sq. In.	Elongation, Per Cent.	Contraction, Per Cent.	Fremont	Bend, Degrees	Treatment	Fig. No.
1	0.34	0.61	0.86	0.007	0.047	88,100	54,650	24.3	35.6	32.5	140	As received.	70
2	0.34	0.61	0.86	0.007	0.047	83,150	55,180	21.0	32.7	30.0	180	900° C. (4) water 680° C. (6) air.	71
3	0.34	0.61	0.85	0.007	0.047	96,600	58,400	22.2	31.4	12.5	180	900° C. (4) air 700° C. (6) air.	72
4	0.37	0.40	0.79	0.008	0.019	83,800	40,000	23.9	32.9	5.0	100	As received.	73
5	0.37	0.40	0.79	0.008	0.019	82,250	49,500	26.7	49.9	10.0	180	900° C. (4) water 680° (6) air.	74
6	0.37	0.40	0.79	0.008	0.019	88,450	49,180	21.4	28.3	6.0	180	900° C. (4) air 700° (6) air.	75
7	0.39	0.41	0.86	0.008	0.019	71,800	35,050	16.8	31.4	6.0	95	As received.	76
8	0.39	0.41	0.86	0.008	0.019	85,850	50,640	23.5	38.7	24.5	180 F.	900° C. (4) water 680° C. (6) air.	77
9	0.39	0.41	0.86	0.008	0.019	83,200	42,810	20.7	29.5	14.0	180	900° C. (4) air 700° C. (6) air.	78
10	0.48	0.41	0.68	0.010	0.019	83,200	38,710	22.6	27.1	7.0	120	As received.	79
11	0.48	0.41	0.68	0.010	0.019	87,500	52,000	24.9	41.9	10.0	180 F.	900° C. (4) water 680° C. (6) air.	80
12	0.48	0.41	0.68	0.010	0.019	90,900	53,600	21.3	29.5	8.5	160	900° C. (4) air 700° C. (6) air.	81

High-manganese Carbon Steel

In March, 1909, it became evident that even with the very best double treatment, the plain carbon cast steels that we had been manufacturing were not good enough for the very severe service to which a great number

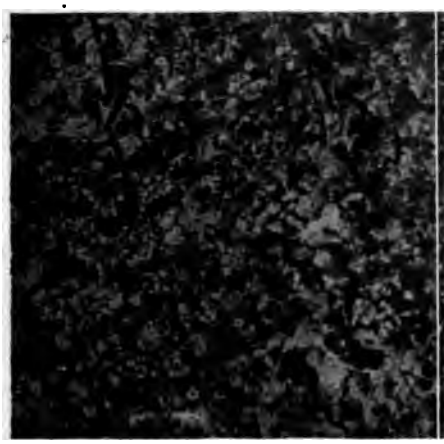

Fig. 79.—0.48 carbon. Electric steel as received.

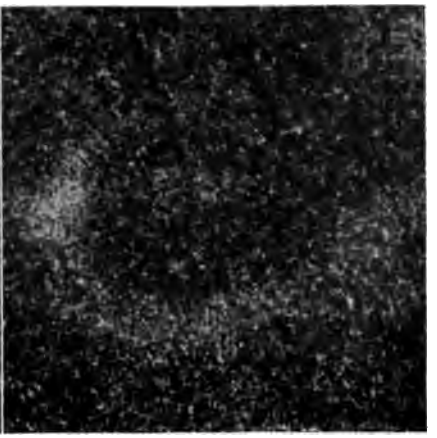

Fig. 80.—0.48 carbon. Same electric steel annealed at 900° C. for 3 hr. and quenched in water. Reheated to 660° C. for 8 hr. and cooled in air.

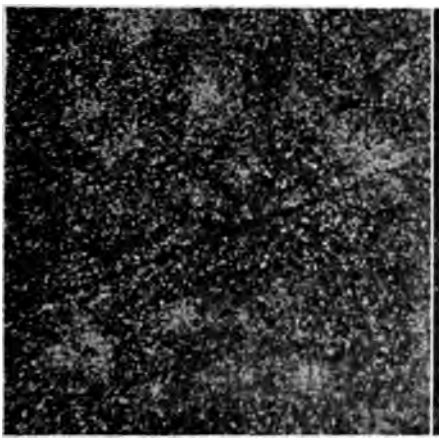

Fig. 81.—0.48 carbon. Same electric steel annealed at 900° C. for 3 hr. and cooled in air. Reheated to 700° C. for 6 hr. and cooled in air.

Fig. 82.—0.24 carbon, 1.33 manganese. Annealed at 900° C. for 4 hr. and quenched in water. Reheated to 640° C. for 6 hr. and cooled in air.

of the castings were subjected. We were at that time manufacturing dredge-bucket castings of carbon steel for the gold dredges of California, Alaska, and other states. Our experience with this type of casting began about 1900, when the gold-dredging business was in its infancy and we had

found that the softer steels, averaging around 0.25 per cent. carbon, wore out and gave way far too rapidly to give satisfaction. In an effort to improve the service of these castings, we had gradually increased the carbon content of the steel; but this led almost immediately to trouble

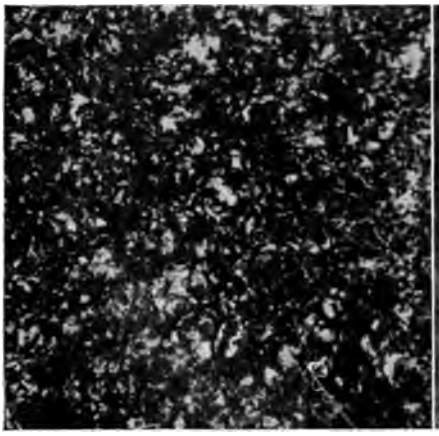

FIG. 83.—0.26 CARBON, 1.76 MANGANESE. ANNEALED AT 1000° C. FOR 20 MIN. AND AIR COOLED. REHEATED TO 750° C. AND COOLED IN AIR.

FIG. 84.—0.26 CARBON, 1.76 MANGANESE. ANNEALED AT 900° C. FOR 4 HR. AND QUENCHED IN WATER. REHEATED TO 680° C. FOR 8 HR. AND COOLED IN AIR.

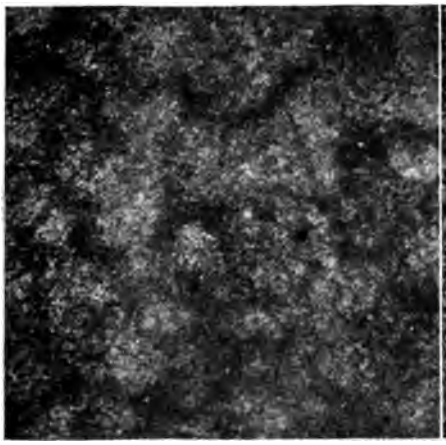

FIG. 85.—0.36 CARBON, 2.42 MANGANESE. ANNEALED AT 900° C. FOR 3 HR. AND COOLED IN AIR. REHEATED TO 660° C. FOR 6 HR. AND SLOWLY COOLED IN FURNACE.

FIG. 86.—NICKEL STEEL; 2.36 NICKEL. ANNEALED AT 900° C. FOR 3 HR. AND COOLED IN AIR. REHEATED TO 650° C. FOR 6 HR. AND COOLED IN AIR.

from breakage before much, if any, wear had taken place. The general shape of these castings, as we were then making them, is shown in Figs. 102, 103, and 104, the latter giving approximate dimensions. These buckets ran in weight from about 800 lb. (363 kg.) to as high as 1600 lb.; and we successfully manufactured, by quenching in water and annealing,

TABLE 5.—*High-manganese Bessemer Steel*

Carbon, Per Cent.	Silicon, Per Cent.	Manganese, Per Cent.	Tensile Strength, Lb. per Sq In.	Elastic Limit, Lb. per Sq. In.	Elongation, Per Cent.	Contraction, Per Cent.	Fremont	Bend, Degrees	Treatment
0.24	0.55	1.33	86,550	57,820	27.1	57.1	14.0	80	900° C. (4) water 630° C. (6) furnace.
0.22	0.49	1.36	81,900	52,150	29.2	54.5	2.5	180 F.	900° C. (4) water 630° C. (6) furnace.
0.22	0.49	1.36	89,450	56,400	26.7	50.7	13.5	180	900° C. (4) water 630° C. (6) furnace.
0.23	0.52	1.36	92,800	61,240	25.2	57.9	6.0	180	900° C. (4) water 610° C. (6) furnace.
0.23	0.52	1.36	88,500	51,750	24.3	45.6	5.0	180	900° C. (4) water 630° C. (6) furnace.
0.24	0.53	1.36	87,200	58,500	26.6	53.5	10.0	180 F.	900° C. (4) water 630° C. (6) furnace.
0.26	0.51	1.36	96,150	65,070	25.6	57.2	15.0	180	900° C. (4) water 640° C. (6) furnace.
0.24	0.52	1.38	92,100	62,990	23.6	49.4	4.0	70 BH.	900° C. (4) water 610° C. (6) furnace.
0.24	0.51	1.38	92,800	59,500	25.3	53.0	19.0	180	900° C. (4) water 660° C. (6) furnace.
0.24	0.49	1.38	91,300	61,670	24.3	53.7	25.5	180	900° C. (4) water 630° C. (6) furnace.
0.24	0.49	1.38	83,450	51,470	27.4	53.5	25.0	180 F.	900° C. (4) water 630° C. (6) furnace.
0.25	0.51	1.38	101,700	72,310	21.6	45.9	23.0	90	900° C. (4) water 630° C. (6) furnace.
0.25	0.51	1.38	93,900	66,460	22.8	52.7	....	180	900° C. (4) water 630° C. (6) furnace.
0.25	0.55	1.38	84,200	51,750	26.8	56.4	12.5	180 F.	900° C. (4) water 630° C. (6) furnace.
0.25	0.62	1.38	96,600	67,830	25.8	52.5	11.0	180	900° C. (4) water 630° C. (6) furnace.
0.26	0.47	1.39	88,800	59,300	27.0	58.7	13.0	180 F.	900° C. (4) water 640° C. (6) furnace.
0.28	0.53	1.40	83,000	52,990	29.6	57.9	27.5	105	900° C. (4) water 660° C. (6) furnace.
0.31	0.51	1.40	84,240	55,300	28.2	52.0	22.0	180 F.	900° C. (4) water 640° C. (6) furnace.
0.23	0.50	1.41	94,900	65,600	21.1	52.9	5.0	180 F.	900° C. (4) water 610° C. (6) furnace.
0.24	0.54	1.41	90,200	55,600	24.6	48.9	12.5	180	900° C. (4) water 610° C. (6) furnace.
0.24	0.54	1.41	86,750	46,590	22.8	33.2	10.0	180	900° C. (4) water 700° C. (6) furnace.
0.27	0.59	1.41	87,900	54,010	26.5	54.2	22.5	180 BH.	900° C. (4) water 640° C. (6) furnace.
0.28	0.62	1.41[b]	92,550	54,650	26.1	53.2	19.0	180 F.	900° C. (4) water 640° C. (6) furnace.
0.22	0.51	1.42	79,400	46,800	28.6	46.2	5.5	180	900° C. (4) water 630° C. (6) furnace.
0.22	0.51	1.42	87,300	57,800	28.8	46.7	10.0	180 F.	900° C. (4) water 610° C. (6) furnace.
0.25	0.55	1.43	92,200	61,850	24.1	53.2	30.0	180 F.	900° C. (4) water 640° C. (6) furnace.
0.25	0.55	1.43	86,600	49,190	24.0	56.7	11.0	180 F.	900° C. (4) water 640° C. (6) furnace.
0.25	0.55	1.44	94,750	65,830	23.3	47.6	10.5	135	900° C. (4) water 630° C. (6) furnace.
0.26	0.51	1.44	91,300	61,400	24.0	46.2	....	180	900° C. (4) water 640° C. (6) furnace.
0.28	0.66	1.45	88,700	52,150	27.0	58.2	26.0	180	900° C. (4) water 640° C. (6) furnace.
0.23	0.48	1.47	92,300	60,970	26.7	53.1	5.0	180	900° C. (4) water 640° C. (6) furnace.
0.28	0.64	1.50	94,000	65,280	28.7	52.3	15.0	145 BH.	900° C. (4) water 640° C. (6) furnace.
0.34	0.53	1.54	99,000	63,900	25.2	41.0	10.0	180	900° C. (4) water 640° C. (6) furnace.
0.26*	0.54	1.76	82,750	56,600	28.2	49.2	30.0	180	900° C. (4) air 700° C. (6) air.
0.26*	0.54	1.76	84,500	57,200	24.7	41.0	30.0	170	900° C. (4) water 700° C. (6) furnace.
0.26	0.54	1.76	75,950	44,020	30.4	50.4	17.5	180	900° C. (4) air 700° C. (6) furnace.
0.26	0.54	1.76	76,590	42,600	29.4	44.6	25.0	180	900° C. (4) air 700° C. (6) furnace.
0.36[b]	0.24	2.42	106,450	54,080	14.0	19.4	5.5	45	900° C. (4) air 700° C. (6) air.
0 36	0.24	2.42	106,000	52,990	13.7	18.3	4.5	50	900° C. (4) air 660° C. (6) air.
0.36	0.24	2.42	97,700	68,800	12.0	20.1	7.5	75	900° C. (4) air 660° C. (6) furnace.
0.36	0.24	2.42	97,950	70,200	17.0	28.1	6.0	70	900° C. (4) air 660° C. (6) furnace.

F = Bent flat. *BH* = Blowhole flaw. * = Laboratory treated. [b] = S 0.073 per cent.. P 0.041 per cent.

TABLE 6.—*High-manganese Electric Steel*

Carbon, Per Cent.	Silicon, Per Cent.	Manganese, Per Cent.	Tensile Strength, Lb. per Sq. In.	Elastic Limit, Lb. per Sq. In.	Elongation, Per Cent.	Contraction, Per Cent.	Fremont	Bend, Degrees	Treatment
0.26	0.38	1.33	80,000	51,500	20.8	31.8	....	...	900° C. (4) air 700° C. (6) air.
0.26	0.38	1.33	109.000	70,000	21.2	28.9	....	...	900° C. (4) water 700° C. (6) air.
0.36	0.33	1.33	85,750	46,100	24.0	35.7	....	...	900° C. (4) air 700° C. (6) air.
0.27	0.28	1.38	96,000	59,000	23.9	31.5	....	...	900° C. (4) air 700° C. (6) air.
0.35	0.33	1.38	84,030	46,790	24.5	35.1	....	...	900° C. (4) air 700° C. (6) air.
0.35	0.33	1.38	106,160	73,960	17.7	34.0	....	...	900° C. (4) water 700° C. (6) air.
0.26	0.38	1.39	73,500	48,500	28.9	43.4	....	...	900° C. (4) air 700° C. (6) air.
0.34	0.33	1.39	75,500	42,000	29.2	50.6	....	...	900° C. (4) air 700° C. (6) air.
0.47	0.39	1.40	106,160	61,380	20.9	36.9	....	...	900° C. (4) air 700° C. (6) air.
0.34	0.51	1.41	78,000	53,000	26.0	34.7	....	...	900° C. (4) air 700° C. (6) air.
0.25	0.24	1.43	82,500	43,500	27.5	47.8	....	...	900° C. (4) air 700° C. (6) air.
0.44	0.33	1.43	94,500	54,500	20.9	34.1	....	180	900° C. (4) air 700° C. (6) air.
0.31	0.42	1.46	77,000	43,000	29.2	43.7	....	...	900° C. (4) air 700° C. (6) air.
0.32	0.47	1.51	78,700	45,500	25.8	41.8	....	...	900° C. (4) air 700° C. (6) air.
0.32	0.47	1.51	78,930	43,950	23.5	31.4	....	...	900° C. (4) air 700° C. (6) air.
0.32	0.47	1.51	80,500	47,290	26.6	35.7	....	...	900° C. (4) air 700° C. (6) air.
0.55	0.24	1.51	91,000	48,000	20.0	33.8	....	...	900° C. (4) air 700° C. (6) air.
0.34	0.65	1.53	94,500	57,500	22.2	37.9	....	140	900° C. (4) air 700° C. (6) air.
0.44	0.51	1.55	97,500	50,500	20.7	34.1	....	...	900° C. (4) air 700° C. (6) air.
0.44	0.51	1.55	99,500	49,000	21.2	34.7	....	...	900° C. (4) air 700° C. (6) air.
0.27	0.36	1.56	98,500	56,750	16.4	20.6	....	...	900° C. (4) air 700° C. (6) air.
0.30	0.47	1.56	84,500	46,780	20.5	33.1	....	...	900° C. (4) air 700° C. (6) air.
0.30	0.47	1.56	90,500	53,500	18.2	29.9	....	...	900° C. (4) air 700° C. (6) air.
0.30	0.38	1.64	76,000	43,000	30.8	57.3	....	...	900° C. (4) air 700° C. (6) air.
0.38	0.24	1.72	100,500	50,000	19.8	34.7	....	...	900° C. (4) air 700° C. (6) air.
0.30	....	1.73	88,400	48,850	23.4	32.0	....	...	900° C. (4) air 700° C. (6) air.
0.34	0.53	1.73	112,500	62,900	18.8	34.7	....	30	900° C. (4) air 700° C. (6) air.
0.34	0.53	1.73	116,500	63,900	18.8	35.7	....	50	900° C. (4) air 700° C. (6) air.
0.37	0.24	1.74	90,000	49,500	15.0	29.9	....	180	900° C. (4) air 700° C. (6) air.
0.37	0.24	1.74	89,000	48,500	16.2	35.7	....	180	900° C. (4) air 700° C. (6) air.
0.47	0.23	1.75	128,000	61,500	10.2	12.3	....	30	900° C. (4) air 700° C. (6) air.
0.39	0.21	1.76	69,000	47, 00	17.5	29.2	....	145	900° C. (4) air 700° C. (6) air.
0.34	0.26	1.79	84,250	45,500	27.2	52.5	....	...	900° C. (4) air 700° C. (6) air.
0.34	0.26	1.79	81,760	46,500	21.2	36.3	....	120+	900° C. (4) air 700° C. (6) air.
0.34	0.26	1.79	95,850	52,000	20.0	32.7	....	120+	900° C. (4) air 700° C. (6) air.
0.27	0.30	1.91	91,400	54,250	23.7	31.2	....	120+	900° C. (4) air 700° C. (6) air.
0.27	0.30	1.91	90,700	52,500	18.8	26.2	....	70	900° C. (4) air 700° C. (6) air.
0.27	0.30	1.91	91,500	52,700	18.7	22.4	....	120+	900° C. (4) air 700° C. (6) air.
0.27	0.30	1.91	91,850	53,700	22.5	35.7	....	58	900° C. (4) air 700° C. (6) air.
0.27	0.30	1.91	93,250	54,000	28.8	31.2	....	110	900° C. (4) air 700° C. (6) air.
0.34	0.26	1.91	105,500	63,000	13.7	20.6	....	120+	900° C. (4) air 700° C. (6) air.
0.32	0.19	2.38	103,400	59,390	18.2	29.2	....	...	900° C. (4) air 700° C. (6) air.

the design shown. A later design was also made, in which the sheet-steel hood at first used was changed to an integral part of the casting which, of course, greatly increased its intricacy. Fig. 102 shows very plainly how far these buckets would wear down at the point of maximum stress before breaking and is a representative photograph of the service secured from steel of about 0.50 per cent. carbon, air-cooled, and annealed. The dredge masters found that when these buckets had worn down to about 1½ in. (3.8 cm.) thick, it paid them to remove the buckets from the line, because they were almost sure to break very shortly afterward; and if a bucket broke in such a position as to let the entire line into the pond the job of fishing the buckets out was very long and costly, involving the services of professional divers.

We thus found ourselves on the horns of a dilemma, as the softer steels wore too rapidly and gave way to some extent and the harder steels broke after a comparatively short life. It was evident that we must turn out a steel that had both high strength and great toughness and, if possible, considerable resistance to wear. We experimented with a number of alloy steels, some of which are shown in Figs. 88–101, but none of them seemed ideal for our purpose. After considerable experimenting we chose the steel shown in Fig. 82 as best suited to the service. As indicated in Fig. 82 and in Tables 1, 5, and 6, this steel was heat treated by quenching in water and reheating, and combined very high strength with great toughness. The average analysis sought was carbon 0.25 per cent. and manganese 1.25 per cent.

We perfected our procedure in manufacturing this steel and heat treating the bucket castings between March and June, 1909, and that summer began to ship these buckets on a commercial scale. Before deciding to ship these castings quenched and annealed, we subjected them to very severe tests in our own shop with a view to determining in a practical way whether the great toughness shown by the tests was truly characteristic of the castings themselves and whether we could find indications of cracks resulting from the quenching. A number of the castings were broken up and found to be entirely free from cracks; tests were also made to determine how much residual stress remained in the castings after heat treatment. We had by this time ascertained that after the second heating it was possible to remove the castings from the furnace and cool them in the air without affecting the physical properties, but desired to prove that the resulting stresses in the castings were too slight to be taken seriously. By making careful measurements between punch marks at various points, both before and after sawing through the castings, we demonstrated that the residual stresses were extremely slight and could be disregarded.

By way of demonstrating that these castings possessed far greater toughness than higher carbon castings of similar strength, we tested both

the high-carbon steel castings and our new product by inserting a manganese-steel wedge in the long eye of the bucket shown in Figs. 102 and 103, and driving this wedge home by blows from a heavy drop ball. We found, after breaking up some thirty or forty castings, that with the harder steel the entire back eye was torn out in two or three blows of the drop ball, leaving the casting very much in the condition shown in Fig. 102, whereas the low-carbon high-manganese castings endured from twelve to over twenty blows of the drop ball and the cracks, which were started in the eyes, spread slowly so that we often literally had to tear the casting apart in order to break it. This made a vivid comparison with the tests on the harder steel, where every man in sight had to get out of the way to avoid the flying fragments.

We manufactured these castings for about 5 years; no cracks appeared in service and very few buckets broke on the dredges. We had some trouble in the shops with one or two patterns, but the total loss from castings cracked in treatment was not over fifteen or twenty out of several thousand manufactured. Fig. 103, which was taken in 1911, shows very strikingly how this steel behaved in service. This bucket wore down under the back eye to about $\frac{3}{16}$ in. (5 mm.) thick on one side and a thin edge on the other without breakage. In our files are photographs from other dredges showing similar performances. Even with this steel, however, the wear was too rapid to satisfy our customers, so we were driven to the expedient of inserting plates of manganese steel, about 1 in. thick, to take up the wear along the bottom, especially at the back eye. With these buckets we made an extremely successful record, which was continued until later when we developed the manufacture of buckets made of manganese steel containing about 1.25 per cent. carbon and 12.0 per cent. manganese, which was adopted for the entire bucket casting in order still further to increase the resistance to wear.

In the 10 years that have elapsed since we produced this steel, we have used it for a number of purposes and during the war we applied it successfully in the manufacture of castings of various sorts for gun carriages. Most of the castings we made for ordnance purposes were of basic electric steel, although the analysis and heat treatment were similar to that used in 1909 for dredge buckets. In the manufacture of these gun-carriage castings, we successfully quenched and annealed a great variety of intricate castings and had no cases where the castings cracked as the result of the quenching. Of course, it is not possible to quench very heavy sections successfully, but there are thousands of castings manufactured in the United States to which quenching can be successfully applied when desirable. That this is so is shown by the fact that a considerable number of steel-casting manufacturers are successfully quenching their castings, and we know of one case in which locomotive frames were successfully treated in this manner by taking them from the water as

soon as they had become black and placing them for the second heating in an annealing furnace at about the temperature of the castings when withdrawn from the water.

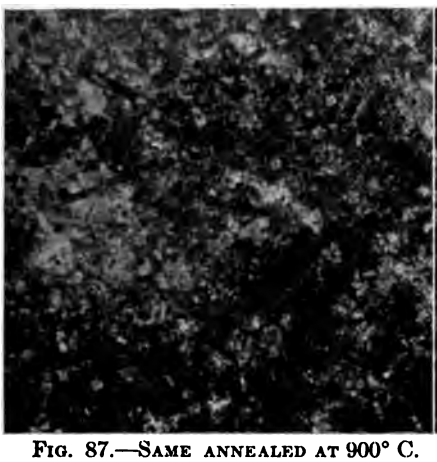

FIG. 87.—SAME ANNEALED AT 900° C. FOR 3 HR. AND QUENCHED IN WATER. REHEATED TO 640° C. FOR 8 HR. AND COOLED IN AIR.

FIG. 88.—NICKEL STEEL: 3.63 NICKEL. ANNEALED AT 900° C. FOR 3 HR. AND SLOWLY COOLED IN FURNACE.

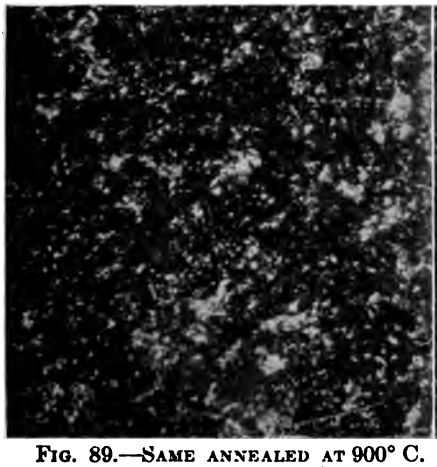

FIG. 89.—SAME ANNEALED AT 900° C. FOR 3 HR. AND COOLED IN AIR. REHEATED TO 710° C. FOR 6 HR. AND COOLED IN AIR.

FIG. 90.—SAME ANNEALED AT 900° C. FOR 3 HR. AND QUENCHED IN WATER. REHEATED TO 680° C. FOR 8 HR. AND COOLED IN AIR.

Tables 5 and 6 and Figs. 83, 84, and 85 are included to show how high the manganese can be carried in Bessemer and electric steel and still give very excellent physical test results. Not so long ago, it was deemed impossible to manufacture serviceable castings that contained, say, over 1.50 per cent. manganese but we contribute herewith analyses

and commercial tests of castings in which we have carried the manganese as high as 2.42 per cent. and still show tests that will pass the No. 2 and No. 3 Army Specifications. We do not claim that the latter percentage of manganese makes the most serviceable castings but we are including

FIG. 91.—NICKEL STEEL; 0.55 CARBON, 3.29 NICKEL. ANNEALED AT 900° C. FOR 3 HR. AND QUENCHED IN WATER. REHEATED TO 680° C. FOR 8 HR. AND COOLED IN AIR.

FIG. 92.—NICKEL STEEL; 0.45 CARBON, 1.27 MANGANESE, 2.42 NICKEL. ANNEALED AT 900° C. FOR 3 HR. AND COOLED IN AIR. REHEATED TO 700° C. FOR 6 HR. AND COOLED IN AIR.

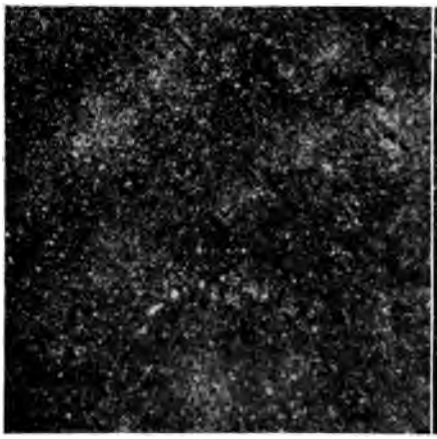

FIG. 93.—NICKEL STEEL; 0.30 CARBON, 1.44 MANGANESE, 2.30 NICKEL. ANNEALED AT 900° C. FOR 3 HR. AND QUENCHED IN WATER. REHEATED TO 680° C. FOR 8 HR. AND COOLED IN AIR.

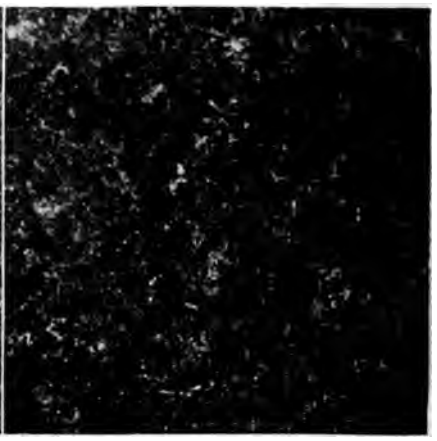

FIG. 94.—CHROME STEEL; 1.23 CHROMIUM. ANNEALED AT 900° C. FOR 3 HR. AND QUENCHED IN WATER. REHEATED TO 680° C. FOR 8 HR. AND COOLED IN AIR.

these data to show that one cannot place sole reliance on physical tests without considering analyses. We also claim that the manganese content combined with proper carbon percentage can successfully be carried a great deal higher than is now generally believed.

Alloy Steels

We have never engaged in the manufacture of cast alloy steels, other than manganese steel on a large scale, so that the tests given in Table 7 and the micrographs, Figs. 86–101, are representative of small lots of

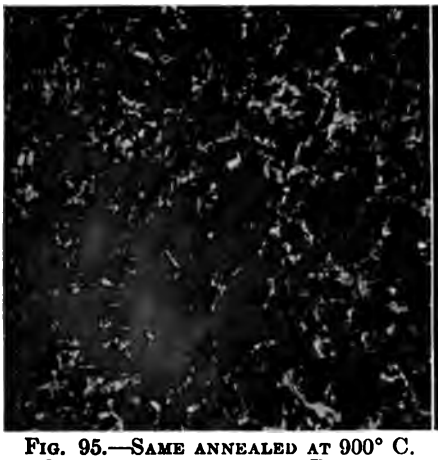

FIG. 95.—SAME ANNEALED AT 900° C. FOR 3 HR. AND COOLED IN AIR. REHEATED TO 700° C. FOR 6 HR. AND COOLED IN AIR.

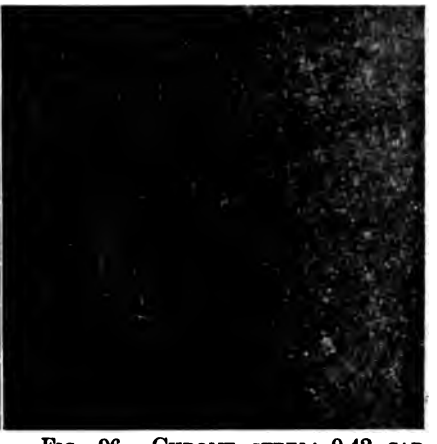

FIG. 96.—CHROME STEEL; 0.42 CARBON, 1.06 MANGANESE, 0.95 CHROMIUM. ANNEALED AT 900° C. FOR 3 HR. AND QUENCHED IN WATER. REHEATED TO 665° C. FOR 8 HR. AND COOLED IN AIR.

FIG. 97.—SAME ANNEALED AT 900° C. FOR 3 HR. AND COOLED IN AIR. REHEATED TO 710° C. FOR 6 HR. AND COOLED IN AIR.

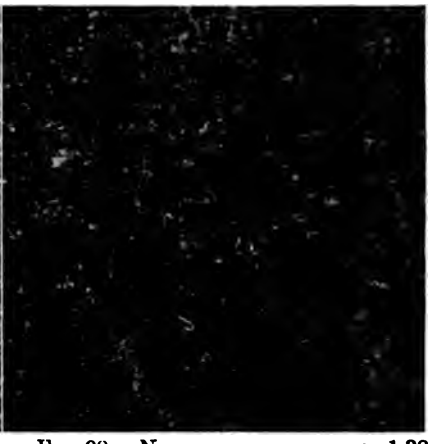

FIG. 98.—NICKEL-CHROME STEEL; 1.38 NICKEL, 0.44 CHROMIUM. ANNEALED AT 900° C. FOR 3 HR. AND QUENCHED IN WATER. REHEATED TO 665° C. FOR 8 HR. AND COOLED IN AIR.

steel made experimentally and heat treated at the laboratory. Reference to the data submitted in Table 7 and to the micrographs readily shows that the double heat treatments we worked out for carbon steels are

TABLE 7.—*Alloy Steels*

Line	Carbon, Per Cent.	Silicon, Per Cent.	Manganese, Per Cent.	Nickel, Per Cent.	Chromium, Per Cent.	Vanadium, Per Cent.	Tensile Strength, Lb. Per Sq. In.	Elastic Limit, Lb. Per Sq. In.	Elongation, Per Cent.	Contraction, Per Cent.	Fremont	Bend, Degrees	Treatment	Fig.
1	0.21	0.36	0.47	2.36	....	....	71,900	47,850	31.3	45.9	25.5	180	900° C. (3) air 650° C. (6) air.	86
2	0.21	0.36	0.47	2.36	....	....	80,700	61,600	24.7	41.9	15.5	180	900° C. (3) water 640° C. (8½) air	87
3	0.24	0.29	0.87	3.63	....	....	84,000	62,990	10.6	16.2	10.0	165	900° C. (3) furnace cool.	88
4	0.24	0.29	0.87	3.63	....	....	91,100	67,200	24.0	22.1	14.0	135	900° C. (3) air 710° C. (6) air.	89
5	0.24	0.29	0.87	3.63	....	....	104,700	73,500	21.7	28.8	2.5	180	900° C. (3) water 680° C. (8) air.	90
6	0.55	0.21	0.90	3.29	....	....	128,600	74,600	6.8	8.2	5.0	50	900° C. (3) water 680° C. (8) air.	91
7	0.45	0.48	1.27	2.42	....	....	116,000	71,650	9.9	16.3	....	45	900° C. (3) air 700° C. (6) air.	92
8	0.30	0.48	1.44	2.30	....	....	104,500	64,220	15.7	25.5	14.0	150	900° C. (3) water 680° C. (8) air.	93
9	0.26	0.39	0.55	....	1.23	....	94,000	72,530	19.7	48.7	34.5	180	900° C. (3) water 680° C. (8) air.	94
10	0.26	0.39	0.55	....	1.23	....	81,600	45,800	24.7	46.2	30.5	180	900° C. (3) air 700° C. (6) air.	95
11	0.42	0.46	1.06	....	0.95	....	107,000	85,100	14.2	19.1	9.5	60	900° C. (3) water 680° C. (8) air.	96
12	0.29	0.53	1.18	....	0.88	....	87,000	51,920	16.7	25.5	12.5	105	900° C. (3) air 710° C. (6) air.	97
13	0.46	0.39	0.77	1.38	0.44	....	106,030	85 770	12.6	19.4	10.0	45	900° C. (3) water 680° C. (8) air.	98
14	0.29	0.37	0.74	2.52	0.84	....	110,640	72,760	9.7	11.9	8.0	55	900° C. (3) water 680° C. (8) air.	99
15	0.37	0.45	0.91	....	1.12	0.34	106,000	87,900	14.4	23.4	6.0	75	900° C. (3) water 680° C. (8) air.	100
16	0.37	0.45	0.91	....	1.12	0.34	95,150	50,870	16.3	22.6	9.0	75	900° C. (3) air 700° C. (6) air.	101

equally applicable to the alloy steels. For instance, Figs. 88, 89, and 90 show a steel of 0.24 per cent. carbon and 3.63 per cent. nickel annealed and heat treated by three different methods. Line 3 of Table 7, Fig. 88, shows this steel annealed at 900° C. for 3 hr. and cooled slowly in the fur-

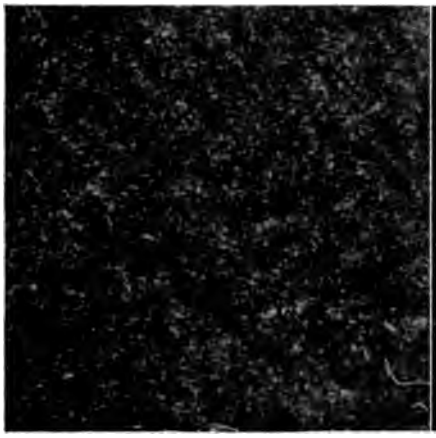

FIG. 99.—NICKEL-CHROME STEEL; 2.52 NICKEL, 0.84 CHROMIUM. ANNEALED AT 900° C. FOR 3 HR. AND QUENCHED IN WATER. REHEATED TO 665° C. FOR 8 HR. AND COOLED IN AIR.

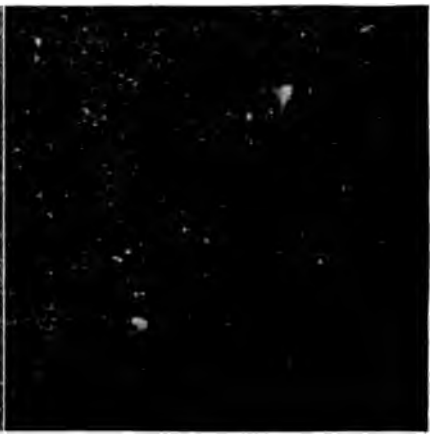

FIG. 100.—CHROME-VANADIUM STEEL; 1.12 CHROMIUM, 0.34 VANADIUM. ANNEALED AT 900° C. FOR 3 HR. AND QUENCHED IN WATER. REHEATED TO 660° C. FOR 7 HR. AND COOLED IN AIR.

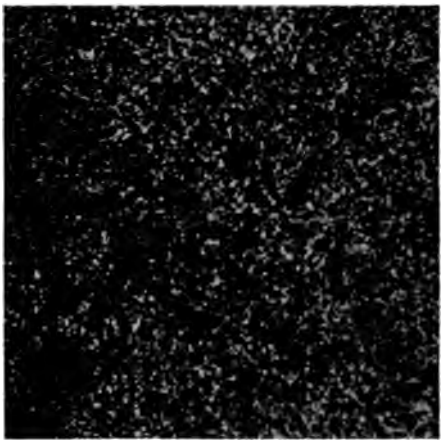

FIG. 101.—SAME ANNEALED AT 900° C. FOR 3 HR. AND COOLED IN AIR. REHEATED TO 700° C. FOR 6 HR. AND COOLED IN AIR.

nace, and lines 4 and 5, Figs. 89 and 90, show the same steel air-cooled and annealed and water-quenched and annealed, respectively. It will be noticed that the tests on the double heat-treated steel are considerably better on the average than the test of the steel with a plain anneal; the micrographs show very clearly why this should be so. The Fremont

tests on these steels, especially in the case of line 5, do not confirm the results of the tensile test; but it should be remembered that in order to get really reliable results from the Fremont test, it is desirable to have several specimens from any one piece of steel and in this case we have but one.

FIG. 102.—BUCKET; 0.50 CARBON, 0.80 MANGANESE. AIR-COOLED AND DRAWN. MAXIMUM WEAR BEFORE BREAKAGE.

FIG. 103.—BUCKET; 0.25 CARBON, 1.25 MANGANESE. WATER-QUENCHED AND DRAWN SHOWING MAXIMUM WEAR BEFORE DISCARDING. NO BREAKAGE.

It does not seem necessary to comment on each of these steels in detail. Figs. 92 and 93, lines 7 and 8 of Table 7, represent a steel made with the idea in mind that some of the nickel in a nickel-steel casting could be replaced by manganese. Figs. 96 and 97, lines 11 and 12 in Table 7, represent steels made with the idea of combining chrome with a high content in manganese. Fig. 94, line 9 in Table 7, gives the properties of a plain chrome steel and is roughly comparable with Figs.

100 and 101, lines 15 and 16 in the table, which gives the properties obtainable from chrome-vanadium cast steel.

We do not consider these tests of alloy steels to be representative of commercial practice but have included the micrographs and tests for what they are worth as a rough comparison with the properties of the steels, which we have made extensively on a commercial scale, and

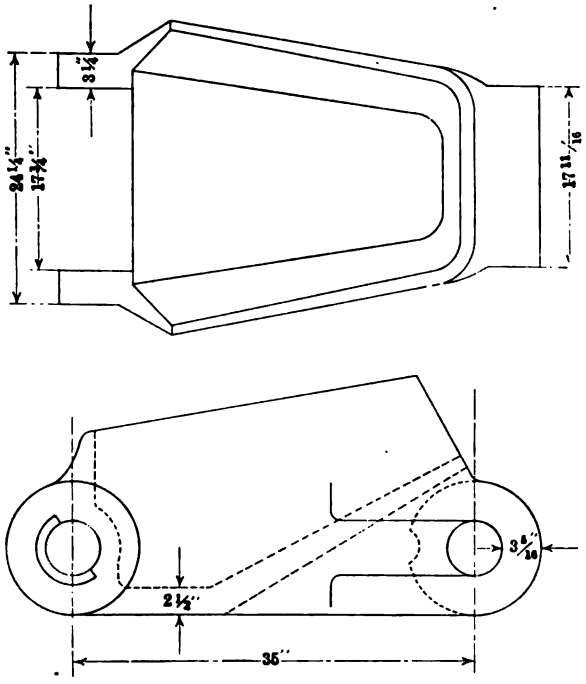

FIG. 104.—UNHOODED BUCKET SHOWING INTRICACY OF CASTINGS THAT WERE SUCCESSFULLY QUENCHED.

incidentally to demonstrate that the benefits of double treatment are not by any means confined to the plain carbon steels. We have always felt that for most of the castings we make, the steel with about 1.25 per cent. manganese has excellent all-around properties, especially when quenched and annealed, and we have preferred to use that steel rather than to go into the manufacture of the more expensive alloy steels. We have been guided in this decision largely by the conviction that the physical properties of the 1.25 per cent. manganese steel are better all around than those that can be obtained from the alloy steels.

In this connection it is worth while perhaps to state that we have made extensive wearing tests over a period of nearly 3 years by means of a small laboratory stone crusher in which several carloads of hard

trap rock were crushed and have in our files some interesting conclusions on the resistance of the different plain carbon, alloy, and commercial manganese steels to this type of wear. Naturally we have been guided in our choice of steels for given purposes by the results of these and other tests.

It is perhaps of interest to call attention to Fig. 82, which represents a steel of the following analysis: carbon 0.24 per cent., silicon 0.53 per cent., manganese 1.33 per cent., copper 0.70 per cent. Annealed, water-quenched, and reheated this steel gives the following test: tensile strength, 92,200 lb. per sq. in.; elastic limit, 66,300 lb. per sq. in.; elongation, 21.9 per cent.; reduction of area, 47.5 per cent.; Fremont, 7.5; bend, 180°. This test is comparable with a test on steel of the same analysis, except for the copper content, which is shown in line 1 of Table 5. Of course, it is not possible to conclude from the results of one test that the higher strength and lower ductility of the copper steel are due to the copper alone. The results may have been affected by slight differences in heat treatment, although the annealing and drawing temperatures in the two cases are identical.

Conclusion

In closing, we again would like to draw attention to the fact that most of the results given in this paper have been obtained from operations on a commercial scale carried out over a period of about 13 years and that almost all of the conclusions arrived at have been confirmed by the examination of many hundreds of tests made on castings annealed on a commercial scale. There are some points on which we do not claim to have arrived at a definite and final conclusion, notably the effect of phosphorus and sulfur on the formation of ferrite ghosts in double treatment, but we thought our experiences of sufficient interest to give them for what they are worth.

Most of the work described in this paper was carried out in close touch with Prof. Henry M. Howe, who acted in the capacity of consulting metallurgist and followed closely the experiments made. The authors wish also to acknowledge their indebtedness to Mr. Clement Le Boutillier who was the chief chemist for many years and supervised the manufacture of the steel; to Mr. A. S. Hummel, metallurgist at the Easton Plant; to Mr. M. F. Apgar who, as superinetndent of the works, contributed largely to our success in putting into practice the treatments determined upon by laboratory research and whose whole-hearted support of our contention that it was safe to quench complicated carbon-steel castings in water contributed very largely to the solution of many of our most difficult problems.

DISCUSSION

H. M. Howe, Bedford Hills, N. Y.—We may clear the matter very easily, at least to a very great extent, if we look at these islands of ferrite that occur in steel castings as simply corresponding to the ferrite ghosts in steel forgings, representing the differentiation that takes place in the solidification of the molten mass. In that solidification you have dendritic crystals forming and in between those dendrites you have concentrated the non-ferrous substances. Some of those substances, notably phosphorus and oxygen, have the power of increasing the quantity of ferrite that forms where those dendritic places persist. The fillings between the dendrites are enriched in phosphorus or oxygen, and are removed by diffusion in annealing extremely slowly. Carbon diffuses itself in steel with astonishing rapidity. The distance it migrates in a very short while is extraordinary, but phosphorus and oxygen apparently move slowly from the place in which they have been concentrated in freezing, so that unless you heat to a very high temperature and for a great length of time, annealing does not have any appreciable effect on the local enrichment in steel castings.

What is the effect of phosphorus or oxygen and how does it come that a local concentration of phosphorus or oxygen has the power to create a greater proportion of ferrite than should be normally present? It seems to me that a simple explanation is that phosphorus or the oxygen shifts the eutectoid point to the right. That is to say, in the presence of phosphorus or oxygen, pearlite has a much higher carbon content than in their absence, with the result that for a given carbon content of the steel as a whole there is a smaller content of pearlite and a larger content of proeutectoid ferrite. Where there is a ghost there is a concentration of oxygen or phosphorus, and this concentration by raising the eutectoid ratio of carbon gives locally a larger proportion of ferrite than exists elsewhere in the mass.

The fact that Mr. Maunsel White discovered high-manganese steel 25 yr. ago has been overlooked to a surprising extent. Mr. Abbott called our attention to it in a paper 6 or 7 yr. ago. It is interesting to look at that manganese as simply representing the equivalent of nickel. A 12 per cent. manganese steel is not far from the equivalent of 25 per cent. nickel steel; 1 per cent. manganese does much more cheaply the work of 2 per cent. nickel. So, if you look at this 1½ per cent. manganese steel as corresponding to a 3 per cent. nickel steel, you wonder that people have passed it by so long.

P. H. Brace, Pittsburgh, Pa.—Doctor Howe, do you believe that the oxygen is in the solid solution? In the laboratory of the Westinghouse Elec. and Mfg. Co., we have done some work on the gases in steels and

find indications that at the high temperatures reactions that generate gases take place.

H. M. HOWE.—I think the phenomena are quite compatible with the idea that it is a case of solid solution of the oxygen, which at high temperatures might react on the carbon. If the oxygen was not in solid solution but there was a foreign body, it is a little hard to see why it should have its specific effect on the properties of the steel.

JOHN H. HALL (author's reply to discussion*).—Mr. Charles Y. Clayton, of the University of Missouri, has called our attention to Figs. 67 and 68, page 372. In his opinion, the dark constituent in those figures is probably troostite and he questions the existence of sorbite in the specimens illustrated. Mr. Clayton sent us photomicrographs very similar to our Fig. 67, in which the dark constituent is undoubtedly troostite, and we believe that our description of Figs. 67 and 68 is in error and that both figures show martensite and troostite.

* Received Jan. 15, 1920.

Effect of Cold-working and Rest on Resistance of Steel to Fatigue under Reversed Stress*

BY H. F. MOORE,† M. M. E., AND W. J. PUTNAM,‡ B. S., URBANA, ILL.

(New York Meeting, February, 1919)

THIS paper gives a preliminary summary of results of tests on the resistance to fatigue under reversed stresses of steel subjected to cold-working and of tests to determine the effect of rest on the endurance of steel under reversed stresses. The tests of cold-worked steel were undertaken as part of an investigation by the National Research Council Committee on Fatigue Phenomena of Metals and were made by Mr. Moore; the tests to determine the effect of rest are part of an investigation now being made by Mr. Putnam. Both series of tests were made in the Materials Testing Laboratory of the University of Illinois. The preliminary results of these tests are reported now because it is believed that test data on fatigue strength should be available at the earliest possible moment; however, it should be clearly kept in mind that this paper gives preliminary results and is intended to encourage discussion rather than to give final conclusions. Before giving the data of the tests, the writers will explain the term "fatigue strength" as used in this paper and will discuss briefly the method used in interpreting the test data of repeated-stress tests.

The term fatigue strength, as applied to a material, may be defined as that maximum fiber stress that will cause fracture of the material after the repetition of any given number of cycles of stress, each cycle involving the variation of stress from a minimum to the above maximum. The fatigue strength, then, varies with the number of repetitions occurring in the lifetime of a structural or machine member and with the range of stress for each cycle of stress. The fatigue strength will be lower for a member in which the stress is completely reversed than for one in which the stress varies from zero to a maximum.

In interpreting test data from fatigue tests under reversed stress, the stresses and corresponding numbers of repetitions required for failure are plotted on logarithmic cross-section paper. In 1910, Professor

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* Report of research under the auspices of the National Research Council.

† Research Professor of Engineering Materials, University of Illinois; Chairman, National Research Council Committee on Fatigue Phenomena of Metals.

‡ Instructor in Theoretical and Applied Mechanics, University of Illinois.

Basquin of Northwestern University pointed out[1] that over a considerable range the relation of fiber stress and number of repetitions required for failure was given by the formula

$$S = \frac{K}{N^q}$$

in which S = intensity of extreme fiber stress, in pounds per square inch;
N = number of repetitions required to produce failure;
K and q = constants, depending on variation of stress, method of loading, etc.

This formula may be written

$$\log S = \log K - q \log N$$

If values of $\log S$ and $\log N$ are plotted on ordinary cross-section paper or values of S and N are plotted on logarithmic cross-section paper the foregoing formula is represented by a straight line.

How far a formula of this form represents the behavior of materials under repeated stress is an unsolved question. If such a formula were true for all values of S and N, the limit of stress for an infinite number of repetitions would be zero. Without taking time to discuss the limits of this formula, it may be pointed out that by its use a very clear graphic comparison of the results of two series of tests on different materials can be made. The relative stress ordinates for any given value of N give the fatigue limits of the materials for that particular length of life; a comparison of the steepness of the lines representing the test results shows whether the two materials have the same relative fatigue strength for different lengths of life.

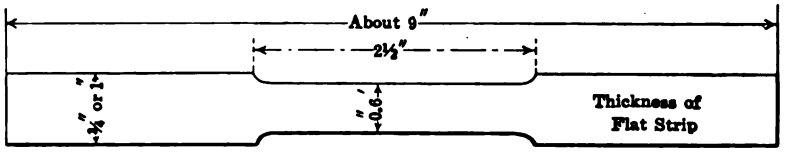
Tests of Cold-worked Steel

It has been known for many years that the elastic limit, the yield point, and, to a less degree, the ultimate strength of steel may be raised by cold-working if a period of rest ensues after the cold-working. There are very few test data on the fatigue strength of cold-worked steel under repeated stress, and tests were undertaken to throw some light on this question whether increased fatigue strength accompanied increased dead-load or static strength due to cold-working of steel.

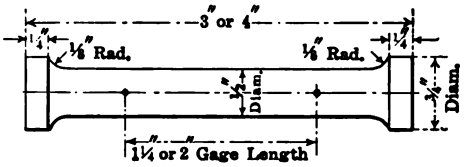
Three series of tests were made: (1) Tests on specimens of hot-rolled steel and on corresponding specimens from the same piece after being cold-worked by stretching with a fiber stress of 57,000 lb. per sq. in. (40.05 kg. per sq. mm.) (corresponding elongation about 0.05 in. per in. of length). (2) Tests on specimens of cold-rolled steel and on corre-

[1] O. H. Basquin: Exponential Law of Endurance Tests. *Proc.* Amer. Soc. Test. Mat. (1910) **10**, 625; see also G. B. Upton and G. W. Lewis: The Fatigue Failure of Metals. *Amer. Machinist* (Oct. 17 and 24, 1912) **37**, 633, 678.

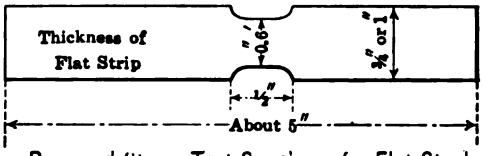
sponding specimens from the same rod annealed to remove the effects of cold-rolling. (3) Tests on annealed specimens of hot-rolled steel and on corresponding specimens from the same rod cold-worked by transverse compression with a fiber stress of 60,000 lb. per sq. in. (42.16 kg. per sq. mm.) (corresponding unit compression about 0.034 in. per in. of thickness).



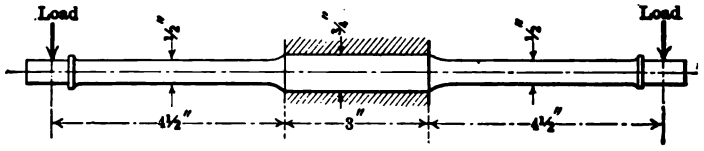
Tension Test Specimen for Flat Steel



Tension Test Specimen for Round Steel



Reversed-Stress Test Specimen for Flat Steel
(Upton-Lewis Machine)



Reversed-Stress Test Specimen for Round Steel
(White-Souther Machine)

FIG. 1.—FORMS OF TEST SPECIMENS USED.

All steel tested had a carbon content of about 0.20 per cent. In series (1), some specimens were tested immediately after cold-stretching, others were tested after cold-stretching and boiling in water for 15 min., and still others were tested after cold-stretching and then resting for 60 days. In series (3), the cold-pressed specimens were tested after cold-pressing, boiling in water for 15 min., and resting about 2 days.

Static tension tests were made in a 100,000-lb. (45,359-kg.) Riehle

testing machine; Fig. 1 shows the shape of the tension test specimen used. In reporting results of static tests, the proportional limit and the American Society for Testing Materials, A. S. T. M., elastic limit are reported. The proportional limit was determined from a stress-strain diagram, and the extensometer used had a gage length of 2 in. (50.8 mm.) and a sensitiveness of reading of 0.00002 in. (0.0005 mm.). The A. S. T. M.

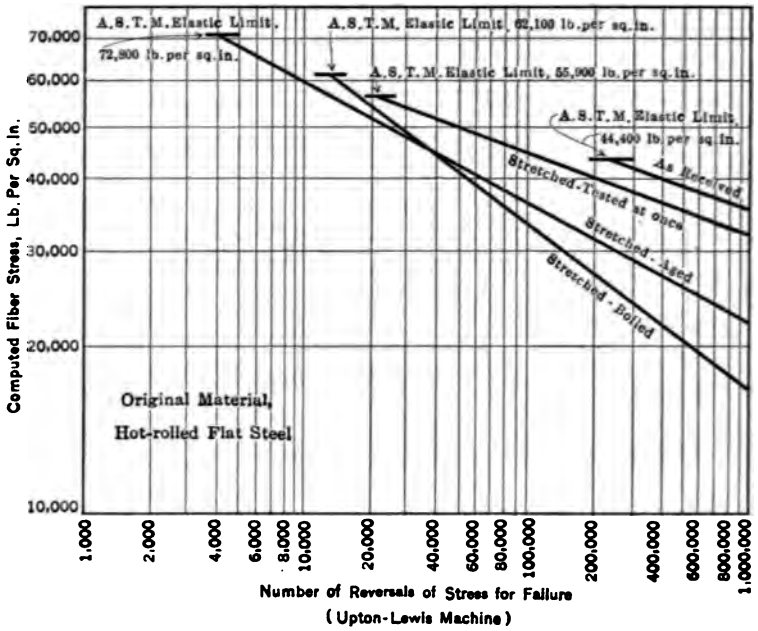


FIG. 2.—TESTS OF COLD-STRETCHED FLAT STEEL.

elastic limit, which is determined by the sudden increase of rate of stretch of the specimen as shown by the extensometer[2] is believed to be a good criterion of the static elastic strength of a metal.

Fatigue tests were made in an Upton-Lewis vibratory testing machine running at 250 vibrations per minute.[3] This machine gives complete reversals of bending stress on a flat specimen. Fig. 1 shows the form of fatigue test specimen used.

Table 1 gives the results of the static tests and Table 2 the results of the fatigue tests. Figs. 2, 3, and 4 show graphically the results of the tests. In plotting the results of the fatigue tests logarithmic coördinates were used. Nominal or computed stresses were laid off as ordinates and numbers of repetitions required to produce failure were laid off as abscissas. The points, plotted on logarithmic paper, formed fairly straight

[2] *Standards*, Amer. Soc. Test. Mat. (1918) 163.

[3] For a description of the Upton-Lewis machine see *Bull.* 98, Engineering Experiment Station, University of Illinois.

lines for each set of fatigue tests, and in Figs. 2, 3, and 4 the inclined straight lines are the most probable straight lines representing the plotted points as determined by the method of least squares.[4]

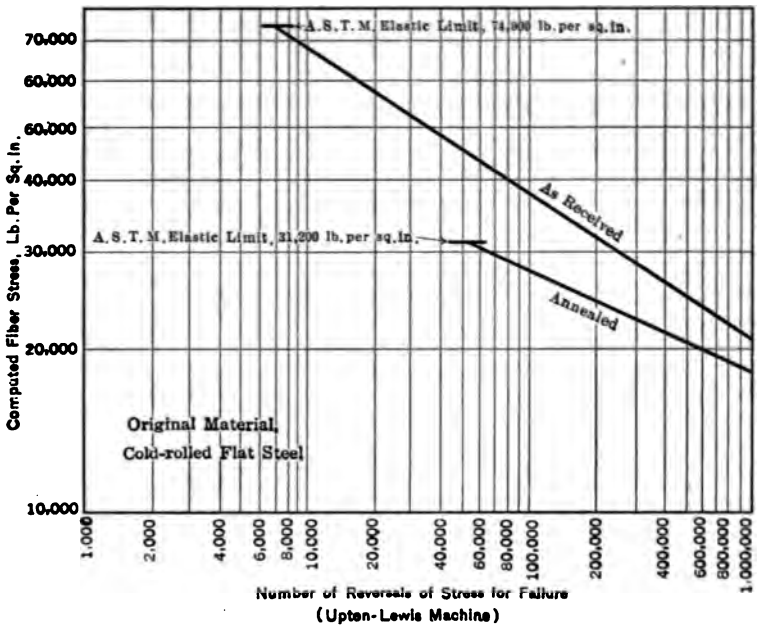


FIG. 3.—TESTS OF COLD-ROLLED FLAT STEEL.

TABLE 1.—*Average Results of Static Tension Tests*

Material	Treatment	Unit Stress, Lb. per Sq. In.		
		Proportional Limit	A. S. T. M. Elastic Limit	Ultimate
Steel from flat strip, hot-rolled	None	40,300	44,400	61,300
Steel from flat strip, hot-rolled	Cold-stretched and tested at once	26,800	55,900	60,300
Steel from flat strip, hot-rolled	Cold-stretched, boiled 15 min.	61,200	62,100	68,700
Steel from flat strip, hot-rolled	Cold-stretched, aged 60 da.	57,200	72,800	77,400
Cold-rolled	None	47,300	74,900	89,900
Cold-rolled	Annealed	27,100	31,200	60,500
Steel from flat strip, hot-rolled	Annealed	30,400	33,300	53,100
Steel from flat strip, hot-rolled	Cold-pressed, boiled 15 min. aged 2 da.	40,500	48,500	62,700

[4] Tuttle: "The Theory of Measurements," 242.

TABLE 2.—*Results of Fatigue Tests*

Specimen Mark	Treatment of Specimen	Nominal Fiber Stress, Lb. per Sq. In. S	Number of Repetitions at Failure N
	Specimens from hot-rolled flat strip		
5	None	40,800	303,600
11		70,900	6,100
16		41,900	288,600
21		79,400	3,400
26		45,700	221,300
32		72,000	8,100
6	Stretched to a fiber stress of 57,000 lb. per sq. in. and tested immediately.	40,500	472,000
9		67,100	5,800
15		41,100	126,800
23		45,400	118,500
25		67,200	4,600
31		62,200	21,200
7	Stretched to a fiber stress of 57,000 lb. per sq. in., boiled 15 min. and tested.	63,700	13,300
12		29,100	226,700
14		76,200	4,800
22		70,500	11,700
28		31,500	112,500
29		22,100	490,500
8	Stretched to a fiber stress of 57,000 lb. per sq. in., aged 60 days and tested.	53,600	30,700
10		44,200	108,600
13		58,000	35,500
24		35,800	120,800
27		31,400	198,200
30		62,200	4,800
	Specimens from cold-rolled strip 1 in. by ¼ in.		
..	Cold-rolled as received.	51,600	36,800
...		51,900	17,100
:::		60,600	10,900
:::.		24,700	588,000
:::::.		37,900	192,700
::::::.		46,800	83,600
.	Annealed from 1450° F.	37,900	23,000
....		35,400	20,400
.....		39,800	10,500
::::		22,200	635,400
:::::		26,900	92,300
::::::		28,800	46,500

TABLE 2.—*Results of Fatigue Tests—(Continued)*

Specimen Mark	Treatment of Specimen	Nominal Fiber Stress, Lb. per Sq. In. S	Number of Repetitions at Failure N
	Specimens from hot-rolled strip ¾ in. by 5/16 in.		
B	Annealed from 1350° F.	77,500	1,600
D		75,000	7,100
F		54,000	26,200
H		47,600	86,800
J		28,900	593,600
A	Annealed from 1350° F, compressed sidewise under a fiber stress of 60,000 lb. per sq. in., boiled, and aged 2 days.	62,500	24,800
C		50,600	157,700
E		40,800	259,000
G		60,700	19,700
I		63,400	6,700
K		66,600	6,000

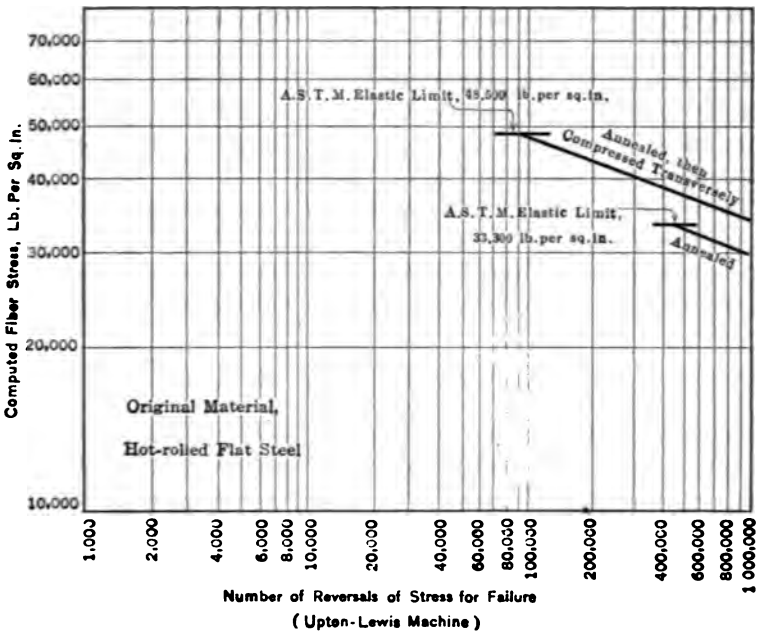


FIG. 4.—TESTS OF FLAT STEEL COMPRESSED TRANSVERSELY.

The A. S. T. M. elastic limit is regarded as the upper limit of elastic strength under any condition and so in Figs. 2, 3, and 4 the inclined straight lines are terminated at their upper ends at an ordinate value representing the A. S. T. M. elastic limit. The values of the A. S. T. M.

elastic limit represent the relative *static elastic strengths* for the different materials and for any given number of repetitions of stress the *fatigue strength* is represented by the ordinate to the inclined line, except for stresses higher than the elastic limit. The inclined lines were determined from the data given in Table 2. The part of the inclined line above the static elastic limit is omitted.

The tests indicate that an increase in the static strength of medium steel due to cold-working does not necessarily indicate increased resistance to fatigue under repeated stress. In interpreting the results of the fatigue tests, it should be borne in mind that if the inclined line representing the test results is steeper for cold-worked material than for material as received or annealed it indicates that for low fiber stresses, and corresponding high numbers of repetitions for failure, the increased fatigue strength due to cold-working diminishes. In the case of cold-stretched material, Fig. 2, for low stresses the fatigue strength of the cold-worked material was found to be actually less than the fatigue strength of the material as received.

For commercial cold-rolled steel, the tests show an apparent approach to equality of strength to resist fatigue between cold-rolled steel and annealed steel as lower stresses are used. It would seem to be rather doubtful whether for working stresses cold-rolled steel would show higher fatigue strength than hot-rolled steel of the same composition. This is also indicated by the relative slopes of the lines for cold-rolled steel and for hot-rolled steel shown in Fig. 5.

The results of tests of cold-pressed steel are interesting in that they show a slight improvement in fatigue strength of steel after cold-pressing as contrasted with diminished fatigue strength after cold-stretching. This result should be confirmed by further tests before being accepted.

Tests on the Effect of Rest

There have been various positive statements that steel subjected to reversals of stress will have its life materially increased if it is allowed resting periods, and it has been experimentally demonstrated by Bauschinger, Hancock, and others that the static elastic strength of overstressed steel is increased by rest. The study of the effect of rest on fatigue strength was the object of the following tests. It was assumed at the start that the rest period should be about such as a machine part would receive in actual practice. Specimens were subjected to repeated stress for one-third of their life, as found by continuous runs to failure on several similar specimens, and were allowed to rest 24 hr. The specimens were then subjected to repeated stress for a second third of their life and again rested for 24 hr. This process was repeated till failure occurred Each specimen tested with resting periods, therefore, had at

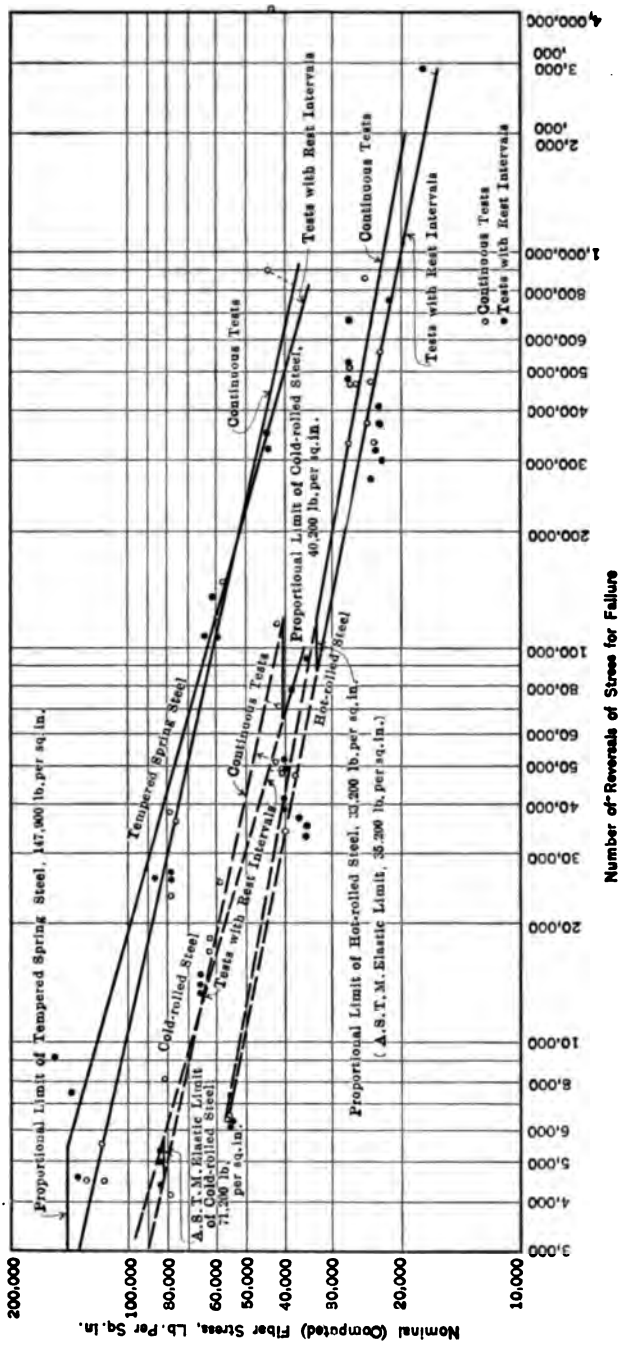


FIG. 5.—FATIGUE TESTS WITH AND WITHOUT REST INTERVALS.

least two rest periods of 24 hr. each. There is an infinite number of other combinations of number and duration of resting periods that could be tried, but it was thought that if rest has a marked effect in increasing life the fact would be evidenced by these tests.

Four series of tests on three different materials were made in the Upton-Lewis vibratory testing machine and four series on four materials were made in a White-Souther rotating testing machine.[5] Static tension tests of the material were made as already described. Table 3 gives the average results of static tests and Table 4 the data for the fatigue tests.

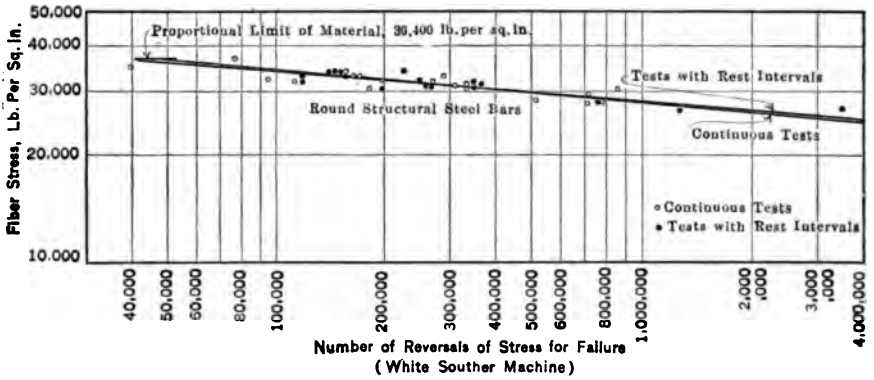


FIG. 6.—RESULTS OF FATIGUE TESTS OF ROUND STRUCTURAL STEEL BARS.

The test specimens used are shown in Fig. 1. Fatigue-test data plotted on logarithmic paper are shown in Figs. 5 to 9. Fig. 5 shows the results of tests made on the Upton-Lewis machine and Figs. 6 to 9 show the results from the White-Souther machine. From the curves no marked benefit of a short period of rest on the fatigue strength of steel can be detected.[6] This is especially true for stresses within the limit of proportionality of the material.

DISCUSSION

JAMES E. HOWARD,* Washington, D. C. (written discussion†).—It is a pleasure to participate in the discussion of a paper on the endurance of steel to repeated alternate stresses. For many purposes the prime and only function of the metal is to resist strains and stresses: stresses of long duration in one direction only; stresses that alternate between a maximum and a minimum in one direction; and stresses of different de-

[5] *Proc.*, Amer. Soc. Test. Mat. (1907) **7**, 616.

[6] The tests on the Upton-Lewis machine show slightly greater fatigue strength under low stresses for the continuous tests. The tests on the White-Souther machine show slightly greater fatigue strength for the tests in which the specimen is given periods of rest.

* Engineer-Physicist, Interstate Commerce Commission. † Received Feb. 8, 1919.

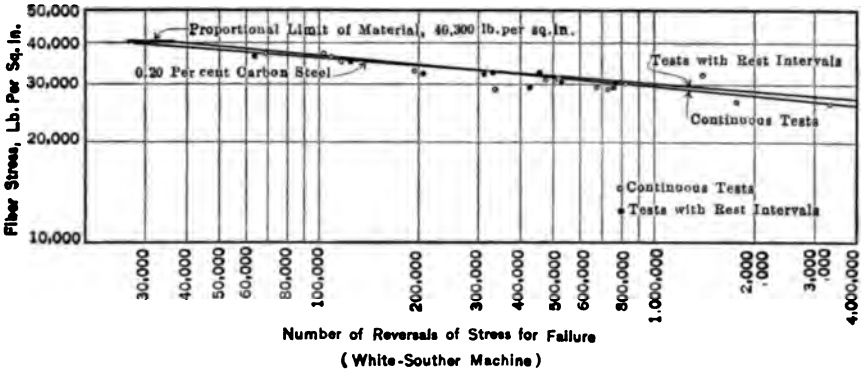


FIG. 7.—RESULTS OF FATIGUE TESTS OF 0.20 PER CENT. CARBON STEEL.

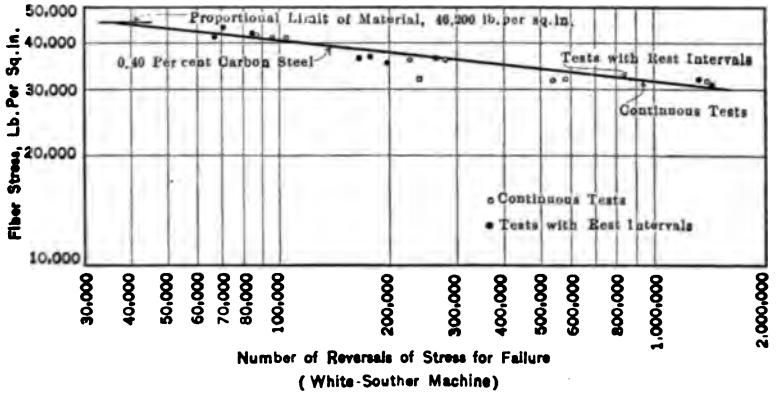


FIG. 8.—RESULTS OF FATIGUE TESTS OF 0.40 PER CENT. CARBON STEEL.

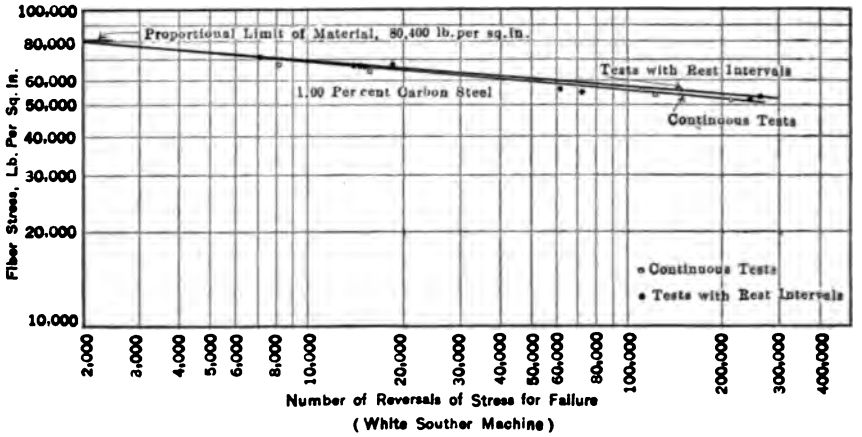


FIG. 9.—RESULTS OF FATIGUE TESTS OF 1 PER CENT. CARBON STEEL.

TABLE 3.—*Average Results for Static Tests on Effect of Rest on Fatigue Strength*

Kind of Steel	Number of Specimens	Proportional Limit, Lb. per Sq. In.	Ultimate Strength, Lb. per Sq. In.	Elongation in 2 in., Per Cent.	Reduction of Area, Per Cent.
Hot-rolled	4	33,200*	54,550	....	64.5
Tempered spring	4	147,900	224,400	3.0	18.2
Cold-rolled	2	40,200†	86,800	13.3	46.3
⅞-in. round structural steel bars	4	36,400	64,800	....	61.3
Nominal 0.20 per cent. carbon steel	17	40,300	64,300	38.8	63.4
Nominal 0.40 per cent. carbon steel	11	46,200	79,100	33.4	57.6
About 1.00 per cent. carbon steel	13	80,400	139,900	1.46	1.49

* A. S. T. M. elastic limit, 35,200 lb. per sq. in.
† A. S. T. M. elastic limit, 71,200 lb. per sq. in.

TABLE 4.—*Results of Fatigue Tests on Effect of Rest on Fatigue Strength*

Continuous to Failure			Resting 24 hr. at One-third and Two-thirds of Continuous Life		
Specimen Mark	Nominal Fiber Stress, Lb. per Sq. In. S	Number of Repetitions at Failure N	Specimen Mark	Nominal Fiber Stress, Lb. per Sq. In. S	Number of Repetitions at Failure N
Hot-rolled steel (Upton-Lewis machine)					
1S1	40,400	40,140	1R1	40,000	41,470
1S2	40,400	48,425	1R2	40,400	39,010
1S3	40,400	49,430	1R3	40,000	52,375
2S1	55,300	6,250	2R1	55,300	6,205
2S2	55,300	6,425	2R2	55,300	6,200
2S3	55,300	6,520	2R3	55,300	7,330
3S1	27,400	501,175	3R1	27,400	462,330
3S2	27,400	468,140	3R2	27,400	524,630
3S3	27,400	330,100	3R3	27,400	670,540
4S1	24,500	371,700	4R1	24,050	270,100
4S2	26,000	465,300	4R2	23,600	315,550
4S3	23,600	330,400	4R3	22,600	295,350
5S1	24,050	475,500	5R1	23,070	369,200
5S2	23,070	558,150	5R2	23,070	410,300
5S3	25,000	852,050	5R3	21,650	750,100
6S1	40,000	42,100	6R1	37,000	36,600
6S2	40,000	33,700	6R2	35,100	35,400
6S3	37,600	47,500	6R3	35,600	33,000

TABLE 4—(*Continued*)

Continuous to Failure			Resting 24 hr. at One-third and Two-thirds of Continuous Life		
Specimen Mark	Nominal Fiber Stress. Lb. per Sq. In S	Number of Repetitions at Failure N	Specimen Mark	Nominal Fiber Stress. Lb. per Sq. In. S	Number of Repetitions at Failure N
Tempered spring steel (Upton-Lewis machine)					
1S1	118,500	5,600	1R1	136,000	4,600
1S2	128,500	4,500	1R2	155,500	9,200
1S3	116,000	4,500	1R3	141,000	7,500
2S1	79,100	38,150	2R1	77,800	25,700
2S2	76,600	36,200	2R2	79,100	26,700
2S3	79,100	23,600	2R3	86,500	25,800
3S1	57,400	149,000	3R1	59,200	107,800
3S2	57,800	106,400	3R2	61,100	137,100
3S3	58,800	119,500	3R3	64,000	108,400
4S1	44,200	895,000	4R1	44,200	320,100
4S2	29,400*	178,800	4R2	44,200	347,500
4S3	28,200*	263,800	4R3	43,700	4,027,300
Cold-rolled steel (Upton-Lewis machine)					
1S1	62,500	18,200	1R1	66,100	13,400
1S2	58,700	25,550	1R2	66,100	14,800
1S3	62,500	16,900	1R3	66,100	14,000
2S1	80,800	4,900	2R1	84,400	4,400
2S2	80,800	8,150	2R2	84,400	5,060
2S3	79,000	4,200	2R3	84,400	5,340
3S1	42,200	115,800	3R1	34,900	92,750
3S2	41,300	71,200	3R2	38,500	77,900
3S3	42,200	50,150	3R3	40,400	49,400
½-in. round structural steel bars (White-Souther machine)					
1S1	27,700	1,240,000	4S1	26,850	14,465,400
	28,300	520,000		27,300	3,464,400
2S1	29,400	713,100	5S1	27,900	756,200
(Broke in grip)	29,400	713,100		27,100	1,275,200
3S1	27,900	779,300	9S1	32,000	117,700
	27,900	713,600		30,400	194,900
6S1	30,500	183,000	10S1	31,050	246,200
	30,400	859,700		31,000	352,900
7S1	30,800	308,000	11S1	31,700	247,900
	32,000	266,200		30,800	264,600
8S1	30,500	334,700	12S1	31,900	346,500
	31,000	334,800		31,200	365,000
1S2	37,300	77,500	8S2	34,100	147,700
	34,000	156,700		34,400	225,300
4S2	34,800	40,300	9S2	34,000	152,400
(Bent)	34,800	40,300		33,600	143,000
5S2	32,800	208,700	10S2	32,500	118,100
	32,800	164,600		33,000	118,200
6S2	32,500	95,400	11S2	33,500	118,500
	32,200	113,400		32,700	156,100
7S2	32,900	167,700			
	32,300	197,600			

*Machine out of order.

TABLE 4—(*Continued*)

Continuous to Failure			Resting 24 hr. at One-third and Two-thirds of Continuous Life		
Specimen Mark	Nominal Fiber Stress Lb. per Sq. In. S	Number of Repetitions at Failure N	Specimen Mark	Nominal Fiber Stress Lb. per Sq. In. S	Number of Repetitions at Failure N
Nominal 0.20 per cent. carbon steel (White-Souther machine)					
2-0	31,700	520,400	2-2	32,500	331,900
	32,300	477,200		32,000	203,000
2-1	32,150	501,200	2-3	32,900	458,700
	32,900	196,300		33,200	311,300
2-4	36,800	104,500	2-6	35,500	126,500
	36,000	110,500		36,100	64,300
2-5	37,000	69,500	2-7	38,000*	11,600 bent
	35,400	118,200		38,000*	11,600 bent
2-9	30,800	811,600	2-8	38,000*	4,200 bent
	29,800	667,500		38,000*	4,200 bent
2-A	29,300	720,300	2-B	30,600	782,100
	29,600	338,800		29,300	766,200
2-D	26,800	1,780,500	2-C	30,500	529,900
	26,000	207,174,900		29,800	426,600
2-E	26,650	3,348,500			
	27,900	526,593,800† bent			
Nominal 0.40 per cent. carbon steel (White-Souther machine)					
4-0-S	40,600	66,400	4-4-S	43,600*	2,400 bent
	41,500	95,100		41,700	66,800
4-2-S	41,500	103,800	4-9-S	43,300	84,300
	42,200	85,900		44,000	70,200
4-C-S	38,300*	36,900 bent	4-H-S	36,700	177,900
	28,300*	36,900 bent		36,500	266,100
4-F-S	36,400	279,700	4-I-S	35,600	198,800
	36,200	226,000		36,600	165,700
4-K-S	32,000	547,400	4-N-S	32,000	1,366,200
	31,900	1,398,200		31,600	1,422,900
4-M-S	32,100	589,100			
	32,200	240,000			
Nominal 1.00 per cent. carbon steel (White-Souther machine)					
4-1-H	68,500	8,200	4-5-H	67,500	18,200
	70,500	7,300		67,100	14,700
4-3-H	67,100	14,100	4-6-H	68,000	14,200
	64,000	15,600		67,500	14,900
4-7-H	54,500	122,300	4-A-H	53,600	260,100
	55,600	116,400		53,400	247,800
4-8-H	52,200	216,100	4-B-H	56,600	61,100
		1,699,700 Broke in chuck		54,900	71,800

* Estimated. † Specimen bent, possibly accidentally.

grees in reversed directions. Such stresses represent some of the conditions that steel must meet in service, tests upon which furnish data of unusual interest and importance.

Experimental research has been directed more commonly upon reversed stresses than upon those acting in one direction. The ease with which reversed stresses may be applied to experimental bars has probably influenced the manner of carrying out this class of tests. In

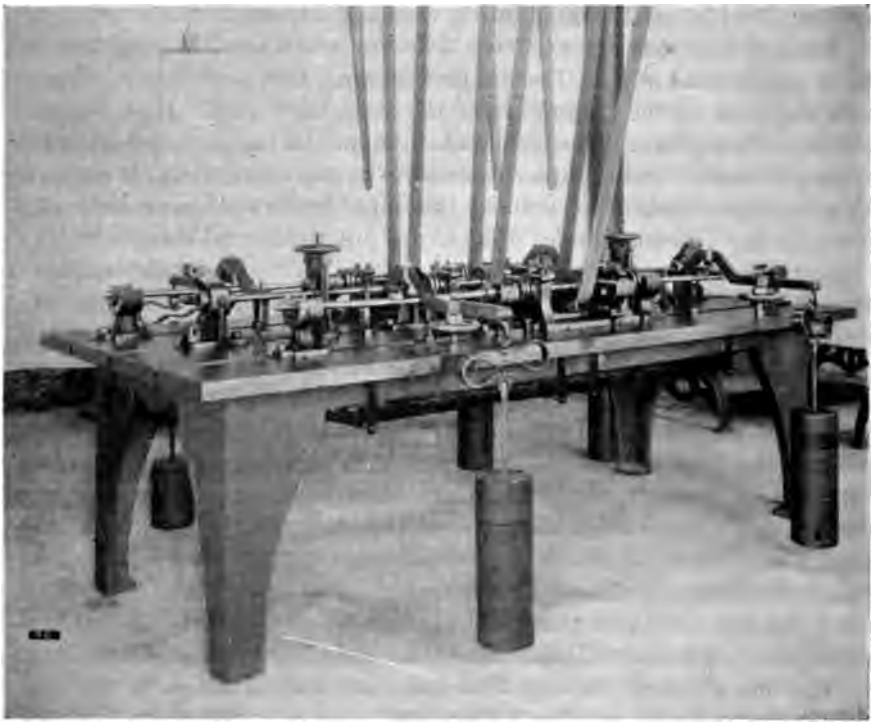


FIG. 10.—REPEATED-STRESS TESTING MACHINE DESIGNED BY J. E. HOWARD, USED IN THE INVESTIGATION TESTS MADE AT THE WATERTOWN LABORATORY, 1892–1908. VIEW OF THE MACHINE WHILE IN OPERATION, FOUR CYLINDRICAL BARS, 1 IN. DIAMETER BY 35 IN. LONG EACH, UNDER TEST. MIDDLE SPINDLE AT REST.

general, it has been convenient to apply reversed stresses in such a manner that the strains of tension and compression have the same values.

Repeated-stress tests, approaching some of the conditions of service experienced by steel, occupy an advanced position in laboratory tests on the physical properties of materials. They aid in showing the absolute and relative values of different steels under certain conditions of service, relations that ordinarily are not established by tests made under specifications governing the acceptance of the material.

The endurance of steel of high stresses, in one direction only, of long

duration, seldom made the subject of direct test, is well illustrated in music wire. In the higher octaves of pianofortes of modern scales, the wires remain under constant tension considerably exceeding 150,000 lb. per sq in. (10,545 kg. per sq. cm.). Fatigue, as commonly understood, does not appear to be known in cases of constant loads acting in one direction, or but few examples have been met. The fractures of music wire usually take place where vibratory waves are interrupted. Without being called upon to do much mechanical work, steel under constant load is capable of enduring very high stresses.

Fatigue fractures are expected to occur when steel is employed in doing mechanical work. The amount of work that a steel bar will perform depends on the magnitude of the straining forces. High stresses lead to early rupture. The results of experimental research indicate that steels will endure certain lower ranges in stresses indefinitely; it remains for direct investigation to establish the upper limits of those zones within which the different grades of steel are practically indestructible.

Notwithstanding the fact that the chief function of most steels is to endure repeated stresses of some kind, very meager information has been experimentally acquired upon the ability of steel to endure the stresses of every-day occurrence. This information is needed to establish rational limits of loads on engineering structures. The primitive properties of steels are modified according to their chemical composition, and also by heat and mechanical treatment. It is generally assumed, but without warrant, that the primitive physical properties, however they may be attained, are indicative of the ability of the steel to endure service stresses of different degrees of intensity and of long or short duration. There are those who place reliance on the aid of the microscope and judge from the microstructure of the steel its ability to meet different service requirements involving strains and stresses.

The one physical constant that has been shown to bear a relation, standing as a limiting value, to repeated stresses is the elastic limit of the steel. It follows by no means, however, that it is safe to approach close to that limit; neither does it appear to have been ascertained by what means the elastic limit and other physical properties shall be acquired in order to best meet prolonged stresses. While the elastic limit is regarded as the upper limit of permissible stress, it is not certain on the one hand that the actual intensity of stress at the point of rupture by fatigue does not exceed that value, while on the other hand there are few, if any, engineering examples of any size in which local strains do not exist in excess of the elastic limit. Methods of test capable of furnishing direct data upon the fitness of the steel to endure service conditions must perforce include features to which the steel is exposed in service, and it may be permissible to add that probably reliable data must be obtained in this manner. The last part of the preceding sentence is

very sweeping since it carries the idea that tests adequate to show the ability of the steel to endure service stresses must be made in kind to the conditions that prevail in service.

In conducting tests having such ultimate objects in view, it is desirable to acquire information upon the phases through which the steel passes during exposure to stresses that finally result in rupture. Such considerations lead to a choice of apparatus in making the tests. Tests on repeated stresses conducted by the writer at the Watertown laboratory were inaugurated in 1888, impressing into service temporary apparatus. It was decided to use cylindrical shafts measuring 1 in. (2.5 cm.) in diameter by 33 in. (82.5 cm.) long in order to have a test bar of substantial size. For a certain length of time the test bars were loaded on a single bearing at the middle of their length. Later, an appropriation was available for a repeated-stress machine, designed by the writer, which was built and put into operation in 1892. This machine, shown in the accompanying illustration, was designed to accommodate four bars under one kind of test and two under another kind.

It was held desirable to use a double bearing for loading the test piece at the middle of its length, in order to apply a uniform bending load over a part of its length. From center to center of the double middle bearing, the distance was 4 in. (10 cm.). The grooved form of tensile test piece was discarded on account of objections to a form of specimen that had no length of minimum section. In a bar subjected to repeated stresses, it was considered even more desirable to have an appreciable length exposed to uniform bending stresses than a length of net section in the case of a tensile test piece. Professor Sondericker, at the Massachusetts Institute of Technology, used 1-in. shafts, in which the loads, as described by Professor Lanza,[7] "were so applied, that a certain portion, greater than 10 in. in length, was subjected to a uniform bending moment." Professor Sondericker's investigations[8] included "the determination of elastic changes, resulting from repeated stresses, and the influence of such changes in producing fracture." Regarding the results, it was stated that "the effect of rest is to decrease the amount of set. In most cases, however, the set lost is soon regained, when the bar is again subjected to repeated stress, especially in the case of the harder steels."

For the Watertown laboratory equipment, a high-speed DeLaval steam turbine was procured, in order to conduct tests at speeds approaching vibratory movements; some 500 or more rotations per second. A rotating shaft appears to afford the only opportunity for conducting tests at high speeds in which acceleration of the mass of the specimen

[7] Gaetano Lanza: "Applied Mechanics," 534. New York, John Wiley & Sons, 1910.

[8] Jerome Sondericker: A Description of Some Repeated Stress Experiments. *Technology Quarterly* (1892) **5**, 70; Repeated Stresses. *Op. cit.* (1899) **12**, 5.

does not interpose as an obstacle to attaining a uniform distribution of the loads. High-speed tests conducted in this manner become introductory to tests on impact, and were so intended. There are two phases to be considered in tests on impact, one having to do with elastic movements not involving permanent sets, and one in which permanent sets are involved and constitute the major effects. The writer's severance of his connection with the Watertown laboratory brought these contemplated tests to a close. The testing apparatus was dismantled soon thereafter and it is believed has not been reassembled. The results of the Watertown tests are included in the reports entitled "Tests of Metals," 1888 to 1908, inclusive. These tests, a little over 400 in number, included cast-iron and wrought-iron bars, bars from muck-bar axle, bars from carbon-steel and nickel-steel ingots, cold-drawn and cold-rolled steel, ordinary hot-rolled steel bars in compositions ranging from about 0.10 to 1.10 per cent. carbon, bars from steel rails, and heat-treated bars. The fiber stresses employed ranged from 10,000 to 60,000 lb. per sq. in. (703 to 4218 kg. per sq. cm.). The tests were made at temperatures ranging from about freezing up to 700° F. (372° C.), with speeds of rotation from 400 to 2200 rev. per min. There were tests that were continuous, that is running the shafts day after day during working hours, and tests in which different periods of rest occurred; tests in which the same fiber stress was used throughout and those in which different fiber stresses were employed on the same shaft. There were also tests on bars that were annealed at intervals and nickel-steel bars, 3 to 5 per cent. nickel and those of about 27 per cent. nickel. Deflections and deflection sets were measured during the progress of the tests. The total number of repetitions of loads on the different bars ranged from a few thousand rotations to upward of 200,000,000. A subsequent examination was made of certain of the bars, after rupture by repeated stresses, to ascertain the effect of repeated stresses upon the properties of the metals, from sections, exposed to the maximum fiber stresses, adjacent to the place of rupture, and comparison bars remote therefrom.

Among the leading thoughts entertained in conducting these repeated stress tests were those concerning the manner in which rupture is approached: Through what phases the metal passes from its primitive state to that of final rupture; what features contribute toward increased life of the metal; whether the effects of repeated stresses are cumulative, eventually reaching that stage in which rupture is certain to take place; whether the effects of earlier stresses can be effaced and the life of the steel prolonged indefinitely by occasional annealing or other treatment whereby the steel is restored to its primitive state; to what extent internal strains are introduced and their manner of action; whether intense internal local strains are set up which eventually lead to the separation of adjacent particles and therefore start the nucleus of rupture; whether the actual

intensity of stress at such incipient places of rupture is represented by the apparent fiber stress carried by the bar, or whether rupture requires a local stress approaching the primitive tensile strength or shearing strength of the steel; whether microscopic evidence is presented of the approach to rupture in the fibers most strained; whether the influence of elongation and contraction of area witnessed in the primitive tensile tests is shown in the endurance of the shafts under repeated stresses, the greater endurance of the shafts at the higher testing temperatures taken in connection with the increased tensile strength of steels at 400° to 600° F. seeming to establish some relation between the primitive tensile strength and the endurance of repeated stresses. These and other thoughts suggested themselves in conducting endurance tests, attaching to matters which seemed involved or relevant to the subject matter. Many of these questions would have to be solved before it would appear feasible to write an equation that would be serviceable in indicating the behavior of steels when exposed to repeated stresses.

Laboratory tests afford opportunities not present in the service use of steels, in respect to control of conditions, whereby complete information is available concerning their endurance of repeated stresses. Elementary features associated with the use of steels in service admit of examination in detail in the laboratory. If features are recognized in the laboratory, which are precursors to rupture, it will give encouragement that a rational method of using steels in service may eventually be arrived at. One of the paradoxes of current engineering practice is the large amount of routine testing that is constantly going on under specifications that are ever changing, the results of which are remotely connected with the inherent value of the material to meet service conditions. Slight consideration has been given the effects of repeated or long-continued stresses under which failures occur, notwithstanding the fact that the profession has been in possession of data of this order for a long term of years.

J. B. KOMMERS,* Madison, Wis. (written discussion†).—So many questions regarding fatigue phenomena remain unanswered that information of the kind given in the present paper is very welcome. The authors are to be commended for the attempt to answer certain questions of the greatest practical importance. Tests on the fatigue of metals require a great deal of time for their successful prosecution, and the investigator knows before he begins that his results will come very slowly. It is evident that if mere cold working has a marked effect on fatigue strength, a very simple method would be available for producing the desired increase in strength. Again, if rest has a marked effect on fatigue strength, this fact could be easily made use of in a practical way.

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† Received Feb. 10, 1919.

It may be well to recall the results obtained by previous experimenters. Eden, Rose and Cunningham,[9] Rogers,[10] Boudouard,[11] J. H. Smith,[12] and Reynolds and Smith,[13] found that annealing steel specimens decreased their fatigue strength. The results of Stead and Richards[14] seem to show that annealing increases the fatigue strength. Rowett found, in some tests on the hysteresis of steel, that for hard-drawn tubes the hysteresis was only about one-eighth of that for the same tubes after annealing. This would imply that the hard-drawn tubes would probably withstand a greater number of reversals for the same applied stress. If, therefore, annealing seems to decrease the fatigue strength, evidently mechanical work during the process of rolling tends to increase the strength.

Referring to Table 2 and Fig. 2, attention should be called to the fact that for all stresses beyond the elastic limit the nominal stresses calculated are stresses that do not actually exist in the specimen. There is considerable evidence showing that the curve determined from these stresses and plotted on logarithmic paper is practically straight to about 1,000,000 reversals. There is also evidence showing that beyond about 1,000,000 reversals the curve tends to bend to the right. The curves drawn in Fig. 2 were all determined from results that were within 490,500 reversals, and for that reason should certainly not be continued beyond 1,000,000 reversals. This illustration makes clear the tediousness of determining such curves for large numbers of reversals. However, they can be correctly determined only by actual tests.

In some tests carried out by the writer on a machine[15] designed by himself and built at the University of Wisconsin, he found that when deflections of cantilever specimens were plotted on logarithmic paper against cycles necessary for rupture, a straight line resulted. These tests were all on specimens stressed beyond their elastic limits. Here again it would be of interest to know whether the straight line determined by these tests holds for deflections and, therefore, stresses within the elastic

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[9] E. M. Eden, W. N. Rose and F. L. Cunningham: The Endurance of Metals. *Proc.* Inst. of Mech. Engrs. (1911) Pts. 3 and 4, 839.

[10] F. Rogers: Heat Treatment and Fatigue of Steel. *Jnl.* Iron and Steel Inst. (1905) **67**, 486.

[11] A. Boudouard: Testing of Metals by the Abatement of Vibratory Movements. *Sci. Abs.* (1911) **14**, 98.

[12] J. H. Smith: Some Experiments on Fatigue of Metals. *Jnl.* Iron and Steel Inst. (1910) **82**, 246.

[13] Osborne Reynolds and J. H. Smith: Throw-testing Machine for Reversals of Mean Stress. *Phil. Trans.* Royal Soc. London (1902) **A 199**, 265.

[14] A. W. Richards and J. E. Stead: Overheated Steel. *Jnl.* Iron and Steel Inst. (1905) **68**, 84.

[15] See J. B. Johnson's "Materials of Construction," 71. New York, John Wiley & Sons, 1919.

limit. If this can be shown to be true, it will be of the greatest interest and importance, because it would make possible the determination of the number of cycles of stress necessary to cause failure at stresses within the elastic limit by the comparatively expeditious tests beyond the elastic limit. The writer is at present designing a machine for tests within the elastic limit, and he hopes to gather some information in an attempt to answer some of the questions just raised.

It may be well to point out that the straight lines determined from deflection-cycle curves are more likely to hold within the elastic limit than the stress-cycle curves, because the deflections actually exist in the specimen, but the stresses calculated by the other method do not actually exist.

A very significant thing in connection with Fig. 2 is that when all the straight lines are extended to the left they tend to coincide to a greater or less extent. This is borne out also by Table 2, in which it is shown that the material as received, with a low elastic limit of 44,400 lb. per sq. in. (3121 kg. per sq. cm.), can withstand 6100 reversals at a nominal stress of 70,900 lb. per sq. in. (4984 kg. per sq. cm.), while the stretched and aged material with a high elastic limit of 72,800 lb. per sq. in. (5117 kg. per sq. cm.) can withstand only 480 reversals at 62,200 lb. per sq. in. (4372 kg. per sq. cm.). If the artificially raised elastic limit of the stretched material has any real effect on fatigue strength, the material should be able to bear a fairly large number of reversals when stressed slightly below its new elastic limit. It cannot do this, which result is very significant.

One reason for this result may be found from experiments on static test specimens. It has been shown[16] that when a piece of wrought iron or steel is subjected to overstrain in tension, it loses its elasticity if immediately reloaded in compression, and vice versa. In the present tests the material was subjected to reversed stresses and a high elastic limit in tension would be of no value in compression. The results in Fig. 2, therefore, tend to show that the artificial raising of the elastic limit is fictitious as far as increased fatigue strength is concerned.

In Fig. 3, the two lines do not converge at the left side of the diagram as in Fig. 2. However, for the cold-rolled material it is seen again that at stresses slightly below the elastic limit the material cannot withstand very many reversals before rupture. These curves are very much like the results obtained by previous experimenters, and the fact that the straight lines converge to the right indicates that the advantage of cold-rolling may be only temporary, as Professor Howe has suggested,[17] and may disappear at large numbers of reversals. It would be very desirable to have such tests carried out considerably beyond 1,000,000 reversals.

[16] Johnson's "Materials of Construction," 604–5, 661.

[17] H. M. Howe: Are the Effects of Simple Overstrain Monotropic? *Proc.* Amer. Soc. Test. Mat. (1914) **14**, Pt. 2, 35.

Comparing Figs. 2 and 3, there is unquestionably a difference. It may be that increasing the elastic limit by cold-rolling has some effect on fatigue strength, but that increasing the elastic limit by tension has none. In this connection I should like to ask whether the figures for ultimate strength after stretching, in Table 1, are based on the original area or on the area after stretching. Fig. 4 shows results such as are usually obtained in comparing annealed and cold-worked steel.

In the matter of effect of rest, Reynolds and Smith[18] found that rest from 3 days to 4 months did not increase the fatigue strength. Eden, Rose, and Cunningham[19] frequently stopped their machine during alternating-stress tests, sometimes for all night. They found that these periods of rest had no appreciable effect. J. H. Smith,[20] in experiments on stress-strain loops, found that during a period of rest steel which had yielded slightly and gave a stress-strain loop tended to approach the primitive condition during a period of rest, but that a comparatively few cycles of stress were sufficient to bring back the original loop. Mason,[21] in some tests of strain measurements during alternating stress, found that rest seemed to improve the material and decrease the strain for a certain stress; but that applications of stress again effaced this improvement. These tests all seem to indicate that if there is any improvement due to a period of rest, the improvement is only temporary and is soon effaced by further application of stress. The results of the present tests bear out the conclusion that very little can be expected from rest in increasing fatigue strength.

In connection with the White-Souther machine, I should like to ask whether the authors encountered any trouble from vibration. In some calculations that the writer has made in connection with the design of a fatigue-testing machine, he found that any eccentricity of the specimen, when the specimen is deflected by weights, sets up serious stresses due to the acceleration effects at high speed.

H. F. MOORE and W. J. PUTNAM (authors' reply to discussion*).—The writers are inclined to question whether it has been experimentally established that steels will endure an indefinite number of repetitions of low range of stress. It is difficult to see how such a conclusion can be reached except by extrapolation from test data, and such extrapolation always gives results of questionable value. From the viewpoint of the designer and the user of steel the writers believe that the exponential formula, which gives results on the safe side of what test data we have, for steel subjected to tens or hundreds of millions of repetitions is a safer

[18] Reynolds and Smith: *Op. cit.* [19] Eden, Rose and Cunningham: *Op. cit.*

[20] J. H. Smith and G. A. Wedgwood: Stress-strain Loops for Steel in the Cyclic State. *Jnl.* Iron and Steel Inst. (No. I, 1915) **91**, 365.

[21] William Mason: Alternating Stress Experiments. *Proc.* Inst. of Mech. Engrs (Jan.–May, 1917) 121.

* Received May 3, 1919.

guide than any endurance limit for an infinite number of repetitions of stress. Whether any statically determined elastic limit can be used as a criterion of fatigue strength is not certain: it is quite certain that the elastic limit as ordinarily determined on steel in its natural condition is not reliable as an index of fatigue strength.

In speaking of the fiber stresses set up in specimens, the term "computed" fiber stress has generally been used. Even when this computed stress is below the elastic limit of the material surface, irregularities and interior flaws may cause localized stress over microscopic areas much higher than the computed stress. These localized stresses are not of importance in determining the deformation of the specimen as a whole, but may be of importance as nuclei for the start of structural damage that will spread under repeated stress.

The writers have made some further study of the test results of Eden, Rose, and Cunningham on the effect of rest on fatigue strength. The results of thirteen series of tests by the above experimenters were carefully plotted on logarithmic paper, one series at a time, and it was found that the lines for tests with rest intervals lay slightly above the corresponding lines for continuous tests. This result is rather to be expected as the rests were accidental rather than planned. The specimens that happened to have the longer "lives" were the ones for which rests were most likely to occur. In the thirteen series there were test records for only five specimens tested with rest intervals that showed a "life" less than 100,000 repetitions of stress, and for these five tests the points lie below the line for the corresponding continuous tests.

Answering the questions of Professor Kommers, the writers would state that the values for ultimate tensile strength given in Table 1 were computed on the basis of the area of the cross-section of the specimen as it was at the beginning of the tension test, that is, on the original cross-section of the specimens tested as received, and for the other specimens on the cross-section after annealing, cold-stretching, or cold-pressing as the case might be.

Concerning vibration of specimens in the White-Souther machine, it may be stated that all specimens tested in this machine were carefully turned all over, and none of them showed marked vibration unless the fiber stress was so high that the yield point of the steel was exceeded, in which case the test was automatically stopped by the bending of the specimen and the consequent disconnecting of the revolution counter. In other series of tests, it has been noticed that a small amount of vibration did not appreciably shorten the endurance of a test specimen as compared with the endurance of other test specimens of the same material for which the vibration was less. If the machine was started or stopped with the weights hanging from the specimen, there was a critical speed at which considerable vibration was evident. This vibration was avoided by releasing the weights while the machine was being started or stopped.

Effect of Time and Low Temperature on Physical Properties of Medium-carbon Steel

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(Chicago Meeting, September, 1919)

THE Youngstown Sheet & Tube Co. has produced a large tonnage of 0.35 to 0.45 carbon forging steel, the acceptance of which was based on the physical properties of test specimens obtained by forging the original 5¾-in. (14.5 cm.) square bloom to a ¾-in. (1.9 cm.) round. Annealed and tempered specimens were required. The test bars were annealed by heating from 900° to 950° C. and cooling slowly in mica. The test bars were tempered by heating to 825° to 875° C., quenching in water, and drawing to 515° to 535° C. Immediately after heat treatment, the test pieces were machined to a diameter of 0.534 in. (1.38 cm.) and tested. Very good physical properties were obtained and no difficulty was experienced in meeting the specifications called for.

	Ultimate Strength, Lb. per Sq. In.	Elongation, Per Cent. in 4 In. Not Less Than
Annealed specimens	78,230 to 92,450	18
Tempered	120,000 to 149,350	9

After experience on this material, production of slightly higher carbon forging steel was started. The method of testing this material required that test blooms be either normalized or annealed. The normalizing consisted in heating to 850° to 900° C. and cooling in still air; the annealing, in heating to 850° to 900° C. and cooling in the furnace or in mica. From the treated bloom, the test pieces were taken either by core drilling or by sawing from points midway between the edges and the intersection of the diagonals. The cores or blocks were turned to a standard 2-in. test piece with a diameter of 0.564 in. and tested without further treatment. The physical requirements of the test pieces were: yield point, pounds per square inch minimum, 44,800; ultimate strength, pounds per square inch, 89,600 to 112,000; elongation, per cent. in 2 in. not less than 17.

Difficulty was encountered as soon as work on this material was started. Many precautions and refinements were instituted but it was found very difficult to secure the required elongation.

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Steel of very good chemical analysis was produced and great care was taken during the heating and rolling. A muffle furnace was equipped with rare-metal pyrometers to secure accurate heat treatment. The test pieces were turned with care so that in testing a straight pull was obtained. None of these precautions seemed to produce the desired results, although occasionally the results were very good. A careful examination disclosed the fact that invariably these occasional results were obtained from test pieces that had been machined the day before they were tested. Mr. I. C. Mackie, chief chemist of the Dominion Iron & Steel Co., said that he had encountered similar conditions with rail steel so that a practice had been developed to rest all rail-steel test specimens for a period of at least 12 hr. before testing. An employee of the metallurgical department of The Youngstown Sheet & Tube Co., who had previously been in charge of a rail mill, reported a similar experience with rail steel. As a result of these experiences and suggestions, all test pieces were rested over night and tested the following day. With this practice no difficulty was experienced in meeting the elongation requirement of the specification. Frequently it was a disadvantage to hold a shipment long enough to afford a 24-hr. rest for the test pieces, so experiments were made that showed that a rest of a few hours at about 120° C. was equal to a rest of 24-hr. at room temperature.

A series of tests was then made to determine the cause for the great difference in elongation and reduction of area shown by core tests made directly after drilling and machining, and core tests made on material that had rested after drilling and machining. While this cause was not found, some interesting data were collected before it was necessary to discontinue the investigation, and it is the purpose of this report to present those data.

In the first series of tests, four cores were drilled from each of six untreated blooms; *i.e.*, blooms direct from the blooming mill. Two cores from each bloom, after machining, were rested two days at room temperature before testing. The two duplicate cores after machining were rested two days on top of a Hoskins hairpin-type electric furnace where the temperature was about 120° C. There are also reported the physical properties of a core from an adjacent piece of the same bloom which had received the heat treatment shown. The machined test piece had rested approximately 12 hr. on top of the Hoskins furnace. The results are given in Table 1.

A second series of tests was made in identically the same manner except that, after machining, a rest of four days was given the test pieces before testing. The results are given in Table 2.

These two series of tests showed that a slight increase in temperature resulted in a great improvement of elongation and reduction of area. There is a general increase in the tensile strength, which, however, was

TABLE 1.—*Results of First Series of Tests*

Heat No.	Yield Point, Lb. per Sq. In.	Ultimate Strength, Lb. per Sq. In.	Elongation, Per Cent. in 2 In.	Reduction of Area	Analysis C	Mn	S	P	Treatment
A-9	61,910	107,200	10.0	12.6	0.53	0.75	0.034	0.050	
	58,370	105,500	11.0	13.4					
Aver...	60,140	106,350	10.5	13.0					Rested 2 days at room temp.
	53,360	109,100	15.5	21.1					
	57,260	109,000	17.0	22.4					
Aver...	55,310	109,050	16.25	21.8					Rested 2 days on furnace (120° C.).
	46,720	88,120	21.0	29.0					Heated at 850° C.; 2 hr., cooled in mica.
B-33	55,850	102,500	11.0	13.4	0.58	0.58	0.033	0.030	
	58,450	103,600	11.0	14.4					
Aver...	57,150	103,050	11.0	13.9					Rested 2 days at room temp.
	59,480	105,200	16.0	18.9					
	60,400	107,000	14.5	16.9					
Aver...	59,940	106,100	15.25	17.9					Rested 2 days on furnace (120° C.).
	45,520	86,550	21.5	29.1					Heated at 850° C.; 1½ hr., cooled in mica.
C-11	54,400	95,150	14.5	18.2	0.49	0.69	0.032	0.011	
	56,570	96,500	14.0	16.7					
Aver...	55,480	95,820	14.25	17.4					Rested 2 days at room temp.
	50,840	100,800	17.0	22.0					
	51,160	97,440	21.0	22.3					
Aver...	51,000	99,120	19.0	22.2					Rested 2 days on furnace (120° C.).
	50,720	94,000	24.0	38.3					Heated at 850° C.; 1½ hr., cooled under cover.
C-30	55,040	95,800	14.5	16.3	0.50	0.66	0.036	0.039	
	53,800	95,520	12.5	17.6					
Aver...	54,420	95,660	13.5	16.9					Rested 2 days at room temp.
	56,450	101,600	19.0	26.4					
	59,320	98,400	18.5	28.1					
Aver...	57,880	100,000	18.75	27.2					Rested 2 days on furnace (120° C.).
	50,250	94,160	24.0	37.7					Heated at 850° C.; 1½ hr., cooled under cover.
D-30	55,240	105,100	7.5	9.4	0.55	0.75	0.031	0.009	
	57,850	104,400	8.0	10.0					
Aver...	56,540	104,750	7.75	9.7					Rested 2 days at room temp.
	59,260	107,300	15.0	20.2					
	58,360	107,500	11.0	13.3					
Aver...	58,810	107,400	13.0	16.7					Rested 2 days on furnace (120° C.).
	46,450	89,760	17.0	21.4					Heated at 870° C.; 1¾ hr., cooled in mica.

TABLE 2.—*Results of Second Series of Tests*

Heat No.	Yield Point, Lb. per Sq. In.	Ultimate Strength, Lb. per Sq. In.	Elongation, Per Cent. in 2 In.	Reduction of Area	Analysis C	Mn	S	P	Treatment
E-8	52,880	101,000	10.5	11.4	0.50	0.67	0.040	0.023	
	49,060	101,500	9.5	10.0					
Aver...	50,970	101,250	10.0	10.7					Rested 4 days at room temp.
	52,080	103,000	17.0	24.2					
	51,110	103,400	17.0	25.8					
Aver...	51,590	103,200	17.0	25.0					Rested 4 days on furnace (120° C.).
	49,720	97,000	22.5	38.3					Heated at 850° C.; 1½ hr., cooled in air.
E-25	51,160	93,400	16.0	22.7	0.49	0.55	0.037	0.017	
	51,640	93,240	16.0	11.3					
Aver...	51,400	93,320	16.0	17.0					Rested 4 days at room temp.
	49,360	93,260	22.0	33.9					
	49,320	94,180	19.0	30.6					
Aver...	49,340	93,720	20.5	32.3					Rested 4 days on furnace (120° C.).
	49,280	93,260	23.5	37.5					Heated at 850° C.; 1½ hr., cooled in air.
F-10	51,560	97,680	11.5	13.0	0.50	0.64	0.037	0.018	
	51,760	97,140	12.0	14.0					
Aver...	51,660	97,410	11.75	13.5					Rested 4 days at room temp.
	49,560	97,520	17.5	24.2					
	50,880	100,500	16.0	24.5					
Aver...	50,220	99,010	16.75	24.4					Rested 4 days on furnace (120° C.).
	48,640	94,760	24.0	39.7					Heated at 850° C.; 1½ hr., cooled in air.
F-30	51,080	96,920	12.0	13.4	0.49	0.60	0.039	0.018	
	49,640	96,400	11.0	13.7					
	50,360	96,660	11.5	13.6					Rested 4 days at room temp.
	50,600	98,400	16.0	21.1					
	53,440	98,480	17.5	22.7					
Aver...	52,020	98,440	16.75	21.9					Rested 4 days on furnace (120° C.).
	53,050	92,500	25.0	39.0					Heated at 850° C.; 1½ hr., cooled in air.

greater in the case of a two days' rest than in the case of a four days' rest.

From adjacent blocks of one bloom, eight core tests were taken and machined at different rates to determine the effect of the speed of machining. Four of the cores were machined at the usual rate and four with

TABLE 3.—*Tests on Adjacent Blocks of One Bloom*

Heat No.	Yield Point, Lb. per Sq. In.	Ultimate Strength, Lb. per Sq. In.	Elongation, Per Cent. in 2 In.	Reduction of Area	Analysis				Treatment
					C	Mn	S	P	
G-10	53,600	93,700	18.5	27.9	0.49	0.66	0.041	0.014	
	54,300	94,000	18.5	25.8					
Aver...	53,950	93,850	18.5	26.8					Machined as usual, tested immediately.
	54,510	94,080	23.0	44.1					
	53,560	93,800	24.0	39.7					
Aver...	54,030	93,940	23.5	41.9					Machined as usual, rested on furnace 20 hr.
	55,400	93,500	19.0	26.8					
	54,700	94,300	18.5	26.6					
Aver...	55,050	93,900	18.75	26.7					Machined slowly, tested immediately.
	54,160	93,260	24.0	40.0					
	53,920	92,640	23.5	39.7					
Aver...	54,040	92,950	23.75	39.8					Machined slowly, rested on furnace 20 hr.

very light cuts and at very slow speed. Two of the cores machined as usual and two of the cores machined very slowly were pulled immediately after the machining operation. The four other cores were rested 20 hr. on top of the electric furnace and then tested. The results, given in Table 3, show no difference in elongation, reduction of area, and tensile strength, between the pieces that were machined as usual and those that were very slowly machined. This indicated that the rate of machining exerts practically no influence on the physical properties.

Sixteen core tests were taken from four adjacent blocks of one bloom. Six cores were machined immediately after core drilling. Two of these six cores were tested immediately after machining; two were rested on the furnace 25 hr. before testing, and two were rested 100 hr. on the furnace before testing. Two rough cores were held four days, then heated for 2 hr. in the tin bath at from 490° to 500° C. They were then machined and tested immediately after machining. Four rough cores were held at room temperature for 16 days. Two of the cores were then machined and tested immediately after machining. The other two cores were heated in the tin bath at 595° to 605° C. for 2 hr. and then machined and tested immediately after machining. Four rough cores were heated to 900° C., held 45 min., cooled in thin mica, and then machined. Two were tested immediately after machining and the other two were rested 25 hr. on the furnace before testing.

The results of the tests made on the 16 cores, given in Table 4, show that the elongation increased with increased length of rest at the tempera-

TABLE 4.—*Tests on Four Adjacent Blocks of One Bloom*

Heat No.	Yield Point, Lb. per Sq. In.	Ultimate Strength, Lb. per Sq. In.	Elongation, Per Cent. in 2 In.	Reduction of Area	Analysis C	Mn	S	P	Treatment
H-11	51,300	96,400	14.5	18.2	0.48	0.66	0.042	0.028	Cores machined immediately, tested immediately after machining.
	49,300	94,800	14.5	18.9					
Aver...	50,300	95,600	14.5	18.5					
	51,410	97,520	21.0	37.2					Cores machined immediately, tested after resting on furnace 25 hr.
	52,170	96,360	21.0	34.5					
Aver...	51,790	96,840	21.0	35.8					
	52,460	93,660	23.0	32.5					Cores machined immediately, tested after resting on furnace 100 hr.
	55,490	92,950	23.0	36.6					
Aver...	53,970	93,300	23.0	34.5					
	56,130	94,080	20.0	26.1					Cores held 4 days, heated in tin pot 490–500° C. 2 hr., machined, and tested immediately.
	60,670	95,900	19.5	26.1					
Aver...	58,400	94,990	19.75	26.1					
	49,600	93,480	21.7	32.0					Cores held 16 days, machined, and tested immediately.
	49,280	94,880	21.0	32.0					
Aver...	49,440	94,180	21.35	32.0					
	50,840	94,440	20.5	31.7					Cores held 16 days, heated in tin pot 595–605° C. 2 hr., machined, and tested immediately.
		94,760	20.0	30.1					
Aver...	50,840	94,600	20.25	30.9					
	48,000	91,300	22.0	35.5					Core annealed 900° C., 45 min. Cooled in thin mica, tested immediately after machining.
	47,300	90,400	22.5	35.2					
Aver...	47,650	90,850	22.25	35.4					
	47,840	89,720	23.5	35.7					Core annealed 900° C., 45 min. machined, rested 25 hr. on furnace.
	49,260	90,800	23.5	36.0					
Aver...	48,550	90,260	23.5	35.8					

ture of the top of the furnace. A rest of 16 days at room temperature increased the elongation and reduction of area about the same extent as a rest of 25 hr. on the electric furnace. This comparison is not entirely accurate, for the tests that were rested 16 days were machined after the rest while the tests that were rested 25 hr. had been machined before the rest. Previously reported tests have indicated, however, that the effect of the machining operation is slight, so the comparison is justified. The tests that were heated to 500° and 600° C. after rests of 4 and 16 days, respectively, and then tested immediately after machining, gave lower elongation and reduction of area values than tests that had been rested at lower temperatures. This is difficult to explain, as it is not

TABLE 5.—*Tests on Cores 207 Days Old*

Cores Rested, Then Machined and Tested Immediately					Original Test Made Shortly After Drilling and Machining		
Yield Point, Lb. per Sq. In.	Ultimate Strength, Lb. per Sq. In.	Elongation, Per Cent. in 2 In.	Reduction of Area	Number of Days Cores Rested	Ultimate Strength, Lb. per Sq. In.	Elongation, Per Cent. in 2 In.	Reduction of Area
52,320	102,600	19.0	30.6	208	101,100	13.5	14.7
54,480	102,600	19.0	30.6	208	99,900	13.5	15.7
.....	103,300	19.0	29.5	208	101,900	14.0	17.0
54,640	101,300	21.5	35.2	207	101,300	16.5	19.9
54,720	101,600	21.5	36.4	207	102,600	15.0	17.3

TABLE 6.—*Tests on Cores 1 to 2 Months Old*

Cores Machined Immediately After Drilling, Tested After Resting at Room Temperature					Original Test Made Shortly After Drilling and Machining	
Yield Point, Lb. per Sq. In.	Ultimate Strength, Lb. per Sq. In.	Elongation, Per Cent. in 2 In.	Reduction of Area	Number of Days Cores Rested	Ultimate Strength, Lb. per Sq. In.	Elongation, Per Cent. in 2 In.
51,420	106,700	21.0	31.4	68	101,840	11.0
52,320	96,720	20.5	28.5	47	93,850	18.0
50,550	93,010	23.0	37.2	50	92,350	16.0
51,880	96,960	21.0	28.7	41	96,640	15.0
57,160	91,030	25.5	42.4	41	90,760	20.0
52,960	94,600	25.0	39.7	41	89,720	17.0
52,840	89,480	25.5	41.9	36	91,810	21.0
51,920	88,600	25.0	37.2	41	92,000	21.5
56,540	96,270	23.0	33.2	40	94,500	16.0
49,780	89,240	24.0	36.7	50	87,750	21.0
53,720	94,540	22.0	30.9	62	94,040	14.5

considered that stresses are induced in the test piece by the air-cooling from temperatures as low as 500° and 600° C. The tests on the annealed cores that were rested on the furnace for 25 hr. after machining gave slightly better elongation values than the tests pulled immediately after machining. The difference is slight and supports the opinion that the influence of the usual machining operation is very little.

Five cores that had remained in the laboratory for 207 and 208 days at normal room temperature were machined and tested immediately after machining. The results obtained from these tests are given in comparison with the results obtained on cores from the same blooms that had been tested soon after core drilling and machining. These results, given in Table 5, show a decided improvement in elongation and reduction of area.

A number of machined test pieces had remained in the laboratory from

36 to 68 days. These test pieces were tested and the results obtained compared with the results obtained on tests from the same blooms that had been tested soon after the drilling and machining operations. The exact data on earlier tests were not available, as the tests were a part of the routine testing and the times drilled, machined, and pulled were not recorded. These results given in Table 6 also indicate a considerable improvement in the elongation for the rested test pieces.

Ingots of the same size and chemical composition as those used in the foregoing tests were rolled on the blooming mill and immediately rolled on the continuous mill to 1⅞-in. (4.7-cm.) square billets. From these billets standard test pieces were turned. Some specimens were tested immediately after machining operation, others were tested after resting for different lengths of time at room temperature, still other duplicate sets were tested after resting at slightly elevated temperatures for the same lengths of time as those at room temperature. Identical results were obtained from the three sets of tests. The average result of tests pulled immediately after machining, the average result of tests pulled after different lengths of rest at room temperature, and the average result of tests pulled after different lengths of rests of slightly elevated temperatures were in close agreement with one another. This series of tests indicated that the size of the finished rolled product, bloom, or billet has an important effect on the physical properties of test specimens taken from that product and tested with no subsequent treatment except resting.

Summary

The tests made on blooms indicate that the poor elongation of the tests made immediately after the drilling and the machining is not due to the machining operation to any great extent.

The ductility of the steel expressed by elongation is greatly improved by some equilibrium adjustment, which takes place slowly at room temperature and much more rapidly at slightly increased temperature.

It seems possible that the difference in the results obtained on tests taken from blooms and on tests taken from small billets may be due to a combination of solidification and rolling strains.

DISCUSSION

F. C. Langenberg,* Watertown, Mass. (written discussion†).—I am inclined to the view that the change in physical properties in the material with which the authors are working is due to the relief of internal strains. The fact that this phenomenon disappeared when the material was rolled into 1⅞-in. square billets is of particular interest. It

* Metallurgist, Watertown Arsenal. † Received Aug. 12, 1919.

is certain that the internal strains set up by cooling in a small billet of this size would be much less than those produced in a billet having a larger cross-section cooling under the same conditions.

The statement that the physical properties of a given test bar were materially changed by standing has been brought to my attention in connection with the testing of gun forgings. I have made a number of observations on forgings produced at Watertown Arsenal to ascertain whether any change in ductility was produced by the aging of the test specimens. The results were always negative, but this can be explained by the fact that a gun forging is generally a forging that has been quenched and subsequently drawn at a relatively high temperature.

In view of the fact that experimental methods for the determination of the resistance of internal strains are available, it would indeed be of the greatest interest if such work could be carried out on the material investigated by Mr. Reinhardt and Mr. Cutler. I think that the authors of this paper should be congratulated particularly in presenting in a concise and clear manner experimental work which proved beyond any reasonable doubt that test specimens may undergo a change in their physical properties upon standing. They have also proved, which is in my mind more important, that all test specimens will not undergo this change.

FRANCIS B. FOLEY,* Washington, D. C. (written discussion†).—Reinhardt and Cutler have shown that test bars taken from certain blooms and broken immediately lack ductility, and that ductility is restored by rest at room temperature for a long time or by heating the bars for shorter periods of time. If these factors work to improve the ductility in the test bars, it is reasonable to suppose that they will also work in the same way to improve the metal in the blooms from which the tests were taken and that the time required for the restoration of ductility will increase with the size of the bloom, so that the smaller the section, the more rapidly this end will be brought about. Along this line of reasoning, we should expect ductility to be restored more quickly in the 1⅞-in. square billets than in the larger blooms. This may in part account for the fact that this phenomenon of brittleness was not noted in the case of the 1⅞-in. square billets but only in the larger blooms. Further, there is the matter of finishing temperature and rate of cooling after work was finished. The smaller sections cool more rapidly in the mill and therefore usually finish at a much lower temperature than the larger sections, and, cooling from a lower temperature, pass through their critical range at a slower rate than if the finishing temperature were high. To balance this there is the tendency to fast cooling due to the smaller section. It is possible,

* Metallurgist, U. S. Bureau of Mines. † Received Aug. 29, 1919.

however, that the rate of cooling through the transformation, by preventing the complete separation of proeutectoid iron to the grain boundaries, may play a part in this brittleness. If the finishing temperature is low and the section small, a small grain size results and we may expect a small grain to give up its proeutectoid iron in a shorter space of time than a large grain. In this respect the small section is favored.

Strains probably also play a part in the production of this brittleness. Bars which show this brittleness (16 to 20 per cent. extension) after 4 days at 120° C., give from 22 to 25 per cent. after cooling in air from 850° C., showing that an air cool from 850° C. does not produce this brittleness in a test bar. Why then should it occur in the blooms? Perhaps because of finishing the blooms at a higher temperature than 850° C., or perhaps the condition may be due to strains set up in the cooling of the large section. There are many factors at work in the cooling of a large section from a high temperature. The grains in the center of the section are likely to be much larger than those near the outside; also, there is a temperature difference between the outside and inside which is probably exaggerated when the bloom starts to cool after finishing. By the time the material has reached its transformation range it is cooling quite rapidly; the outside starts to transform first and commences to lag while the center, still cooling, tends to approach it in temperature. The grains in the outside metal, being smaller than those in the inside, pass through their transformation range more rapidly and are further cooled more rapidly by the air, and so the outside metal soon begins extracting heat from the inside, thus cutting short its period of transformation. We may expect, therefore, a difference between the outside and the center with respect to temperature, grain size, and the arrangement of the microconstituents. These differences are exaggerated by rate of cooling, size of section, and finishing temperature. The size of the section is fixed; the finishing temperature and rate of cooling should be controlled if conditions of this kind are to be prevented. Finishing temperature can be kept low by controlling the temperature of the furnace in which the stock is heated. If the operation is to be a quick one, the temperature at which the work is started can be quite low, whereas if the time consumed in working is long, the starting temperature can be correspondingly higher, the same thing applying where the section is small and the material loses temperature rapidly in the course of the working.

GEO. F. COMSTOCK,* Niagara Falls, N. Y. (written discussion†).— On studying this paper it has occurred to the writer that possibly it was the drilling of the test cores that produced the strains which affected injuriously the authors' unrested bars. They found that the machining had only a slight effect, but did they also investigate the effect of the drilling by trying rest applied to the bloom sections before the cores were

* Metallurgical Engineer, Titanium Alloy Mfg. Co. † Received Sept. 29, 1919.

drilled from them? The writer has heard of a similar effect of rest after machining on test bars from castings, and it would seem more reasonable that the ductility was increased because the rests relieved strains produced in drilling the test cores from the blooms, than because of any peculiar combination of solidification and rolling strains in the blooms which did not exist in smaller billets.

WALTER N. CRAFTS,* Toronto, Canada (written discussion†).—During the forging of 9.2-in. shells for the United States Army, it was noticed that better results were obtained in certain instances when test bars had been held some time before pulling. An investigation, therefore, was made by our metallographer, Mr. F. M. Arnold, to determine the effect of seasoning.

The forging consists of a single operation, piercing a solid cast blank. The test sections were cut from the walls by an oxyacetylene flame, one piece from the nose end and one from the base on the opposite side. The test pieces were not heat treated in any way and the tests and seasonings were carried on at normal atmospheric temperatures. The specifications were as follows: Standard 2-in. test bar, 0.505 in. in diameter; elastic limit, over 19 long tons, or 42,500 lb.; ultimate stress, 39.6 to 49.1 long tons, or 87,500 to 110,000 lb. per sq. in.; elongation, over 13 per cent.

As a preliminary test two heats were selected; BP555, which was tested and reported failed on elongation on Aug. 20; and BO1771, which was tested and reported failed on Aug. 24. On Aug. 29, retests were cut and machined from the same shells and held till Sept. 3 (5 days) when they were pulled. Both sets of results are given below.

	Elastic Limit, Lb. Per Sq. In.	Ultimate Stress, Lb. per Sq. In.	Elongation, Per Cent.	Reduction in Area, Per Cent.
Heat BP555, carbon, 0.54 per cent.; manganese, 0.66 per cent.				
Original bar 1, machined and pulled at once ..	52,410	104,200	10.0	16.0
Original bar 2, machined and pulled at once ..	51,510	104,200	9.5	12.5
Retest bar 1, held 5 days and pulled.....	53,070	105,700	14.0	24.0
Retest bar 2, held 5 days and pulled.....	51,740	106,400	14.0	24.0
Heat BO1771, carbon, 0.55 per cent.; manganese, 0.70 per cent.				
Original bar 1, machined and pulled at once..	52,640	104,000	9.0	9.5
Original bar 2, machined and pulled at once..	51,290	102,800	9.0	11.5
Retest bar 1, held 5 days and pulled.....	52,410	105,700	15.0	23.5
Retest bar 2, held 5 days and pulled.....	53,910	105,900	15.0	22.0

* Assistant General Superintendent, British Forgings, Ltd.

† Received Sept. 23, 1919.

This test showed a marked improvement especially in elongation. A further test was cut out of the last heat (BO1771) after it had lain 12 days after forging. The bars were machined and pulled immediately.

	Elastic Limit, Lb. per Sq. In.	Ultimate Stress, Lb. per Sq. In.	Elongation, Per Cent.
Bar 1, 12 days after forging	52,000	102,400	9.5
Bar 2, 12 days after forging	50,000	102,000	10.0

These results, it will be seen, differ but slightly from the original test, for the conditions are similar throughout, except that the forging had lain dormant with opportunity to season, providing seasoning does occur in steel in the forged condition. They would indicate, therefore, that seasoning did not occur in this forging, at least in the walls of the shell.

It was thought strains might be left in the shell from the forging operation. Accordingly the nose was drilled out of the forging and 3 days later bars were turned up and pulled immediately with decidedly better results.

	Elastic Limit, Lb. per Sq. In.	Ultimate Stress, Lb. per Sq. In.	Elongation, Per Cent.
Bar 1	62,350	105,900	14.5
Bar 2	61,900	105,250	12.5

We then selected other heats in an effort to learn just where this change takes place—whether in the shell itself some days after forging, or in the test bars after machining. Assuming the first case, two heats were selected BP580, which failed that day, and BO1769, which had failed 10 days previously. Three blocks were cut from one shell of each heat, called A, B, and C, and were tested in the manner indicated with the results below:

	Elastic Limit, Lb. per Sq. In.	Ultimate Stress, Lb. per Sq. In.	Elongation, Per Cent.	Reduction in Area, Per Cent.
Heat BP580, carbon, 0.52 per cent.; manganese, 0.74 per cent.				
Original test, Sept. 3	50,150	97,900	10.5	15.5
Bar A, machined Sept. 4 and pulled immediately	50,850	97,450	9.5	15.0
Bar B, machined Sept. 4, held 5 days, pulled Sept. 9	52,850	99,000	12.5	Not reported
Bar C, unturned block held 5 days, machined Sept. 9, and pulled immediately	51,750	102,100	11.5	Not reported
Heat BO1769, carbon, 0.53 per cent.; manganese, 0.75 per cent.				
Original test, Aug. 24	51,500	103,500	10.0	13.0
Bar A, machined Sept. 4 and pulled immediately	48,850	104,300	10.5	13.5
Bar B, machined Sept. 4, held 5 days, and pulled Sept. 9	55,100	104,800	15.5	24.0
Bar C, unturned block held 5 days, machined Sept. 9, and pulled immediately	55,350	105,100	14.0	19.5

The A bar of the second heat, even after 10 days rest in the forged condition, did not show improvement but in both cases the C bar, which had been cut out of the shell at the same time as the others, showed better results. This condition was similar to boring out the nose, for theoretically the strains in these bars were released upon being cut from the shell.

Assuming the second case as a possibility; namely, that seasoning may occur in the test bar after machining, from the last heat (BO1769) there were also cut and machined on Sept. 4 eight other bars, D, E, F, G, H, I, J, and K, which were pulled on succeeding days with fairly consistent results.

Test Bar	Elastic Limit, Lb. per Sq. In.	Ultimate Stress, Lb. per Sq. In.	Elongation, Per Cent.	Reduction in Area, Per Cent.
D pulled Sept. 6	51,050	103,000	13.5	Not reported
E pulled Sept. 7	53,750	103,500	15.0	Not reported
F pulled Sept. 9	50,000	102,200	14.0	22.0
G pulled Sept. 10	51,160	102,600	14.5	20.0
H pulled Sept. 11	52,600	103,800	16.0	23.0
I pulled Sept. 12	54,000	104,850	15.0	22.5
J pulled Sept. 13	51,600	104,900	16.0	20.0
K pulled Oct. 4	50,000	103,850	17.0	23.5

Heat	Carbon, Per Cent.	Manganese Per Cent.	Test	Elastic Limit, Lb. per Sq. In.	Ultimate Stress, Lb. per Sq. In.	Elongation, Per Cent.	Reduction in Area, Per Cent.
BO1567	0.53	0.64	Original test	52,550	102,350	10.5	13.0
			Held 5 days	51,200	102,450	13.5	18.5
BO1581	0.50	0.71	Original test	52,100	102,200	10.0	10.5
			Held 5 days	55,000	105,000	14.0	21.6
BP577	0.59	0.77	Original test	51,400	102,350	12.0	16.0
			Held 5 days	52,000	102,300	14.5	21.5
BP513	0.57	0.73	Original test	52,150	99,500	9.5	10.5
			Held 5 days	53,400	100,000	15.0	23.5
CD169	0.54	0.65	Original test	53,300	97,300	7.5	9.5
			Held 5 days	48,850	102,100	14.0	17.5
BP570	0.53	0.70	Original test	55,950	109,650	10.5	14.5
			Held 5 days	54,000	106,600	13.5	20.9
BP575	0.57	0.67	Original test	53,150	101,550	9.5	9.5
			Held 5 days	52,400	102,550	11.0	17.0
BP526	0.56	0.76	Original test	56,900	107,859	11.0	16.5
			Held 5 days	57,300	109,250	12.0	20.5
BO3638	0.59	0.79	Original test	57,100	113,400	11.5	17.5
			Held 5 days	52,100	100,400	15.5	22.0
BO1750	0.50	0.79	Original test	50,150	98,550	9.5	13.0
			Held 5 days	47,800	98,400	15.0	29.0

These results improved the longer the machined test pieces were held, though 5 days were quite ample for our needs.

Numerous other tests were pulled with such uniform results that it became our practice to hold the machined test bars 5 days whenever possible. The results of some comparative tests are given in an accompanying table.

Our experience indicated that almost without exception the process of seasoning the machined test pieces, while materially increasing the elongation and reduction in area, did not affect either the elastic limit or the ultimate stress to any considerable degree. Many of these bars were subjected to microscopic examination both before and after seasoning and pulling. With the equipment at our command we were unable to discern any change whatever in the microstructure of the steel. The signing of the armistice stopped all work so that we were unable to arrive at any conclusions or even theories.

E. GYBBON SPILSBURY, New York, N. Y. (written discussion*).—In confirmation of Mr. Walter Crafts' remarks, it may be interesting to put on record an experience which I had some years ago at the Trenton Iron Works. We had been buying for some years a very special grade of low-carbon open-hearth steel from the Bethlehem Iron Co.; this steel was especially soft and had to be very uniform for the class of wire for which it was used. One shipment of billets, on being reheated and passed through the rolls, turned out so brittle that the rods made from it were absolutely useless. On our complaint, a very thorough investigation was made, by the Bethlehem Iron Co., of the heat from which this steel was made, and it was found that the charge of pig in the furnace was exactly the same as had been previously used—the same iron ore, the same limestone, and the same coke had been used in the furnace. One of the noteworthy features was that no amount of annealing seemed to soften the rods.

I started a very exhaustive series of analyses to try to discover the cause of the trouble. None of the chemists to whom samples were submitted detected any foreign ingredient in the steel until after a number of experiments Mr. Porter Shimer, of Easton, Pa., found a slight evolution of hydrogen. On investigation, the Bethlehem Co. found that the trouble coincided with the installation of the Archer gas producers in the plant and the use of the gas so produced in this special open-hearth furnace. Another heat was then made in the furnace, using the old form of producer gas, when the resulting material was perfectly satisfactory.

The first lot was so bad that the Bethlehem Iron Co. did not even care to have us send back the billets. About 4 yr. later, it was decided to

* Received Oct. 15, 1919.

use these billets to fill an order for very common rods. After some billets had been run through the mills, the foreman was surprised to find that these rods were as soft as any of the soft Bethlehem steel had ever been. A careful examination showed no trace whatever of hydrogen being left in the steel.

It is very evident therefore that this, while exceptional, is a clear example of the benefit of the aging of steel in enabling gases, such as hydrogen, to gradually escape at normal atmospheric temperatures. That the hydrogen can pass from one portion of the steel, under heated conditions, was well demonstrated by the fact that annealing of the rods drove the hydrogen to the center of the rod; which center then became as brittle as glass while the outside covering of the rod had somewhat softened.

W. E. BUCK, St. Louis, Mo.—During the war, we were engaged on forging 8-in. 9.2, and later on the 9.5 shells. At the beginning of the British contracts, we were allowed no leeway whatever, and when we began to forge from ingots cast in the exact weight, or as near exact weight as possible, we were in trouble; sometimes it was with the elongation, sometimes the pieces were too soft, or the elastic limit would be too low or too high. In February, 1915, a heat that the inspector refused to pass was set out in the yard, where it lay until the following June; the elongation was fair and the elastic limit was a little low. After the period of waiting, the Sandberg process having been adopted, we made new tests of those shells and found an average increase of 4 per cent. in the elongation and enough increase in the elastic limit so that all the test pieces pulled well over. We found the same thing true also in a test piece cut with a core drill about 1 in. in diameter; it would not pull in until it had a chance to age. All the British forgings referred to were made of cast steel.

J. A. BRETZ, Bettendorf, Ia.—We have had some difficulty in the last few months in meeting the ordnance specifications for railroad castings but found that our tests could be greatly improved in elongation, and to a smaller extent in reduction, by the accelerated seasoning or heating. In a number of cases, being compelled to pull tests within 2 or 3 hr. after the pieces had been annealed, we found we could equal the longer seasoning of the 5 days' rest by bringing the temperature up to 200°, or 300°, or even 500° F., without any detrimental effects on the steel; and without any great increase in the tensile strength or in the yield points. We are still experimenting along this line. We assume the steel, as it is used, will have many times 5 days in which to rest and that the accelerated resting period, or heating to 250°, would equal probably 16, 25, or 30 days of rest at ordinary temperature. Along with the resting period, we found a slight increase in the percentage reduction and a con-

sistent but small increase in the elastic limit and ultimate strength of the steel.

THE CHAIRMAN (J. W. RICHARDS, So. Bethlehem, Pa.).—It seems to me that the one moral of this discussion is that we need another term added to the specifications for testing materials: the lapse of time after the test piece is made before it should be tested.

W. N. CRAFTS.—There is a possible explanation for the requirements of the different kinds of steel. The man who is acquainted with electric steel will judge from the appearance of the fracture, the temperature at which the steel has been poured. We do that to a moderate extent. We received in this plant six or eight makes of steel. Two of those makes indicated very hot steel pouring, and we had more difficulty with those but found that they would respond to this seasoning test more than the other steels. That seems to me a possible indication that steel makers need to reduce their temperature. I think one of the fundamental considerations is to keep the temperature of steel down as low as it will pour satisfactorily. This particular investigation, I think, indicates that very strongly.

G. A. REINHARDT (author's reply to discussion*).—Mr. Crafts' observations on shell-forgings parallel ours on square blooms. He was able to carry his investigation a step farther and showed that strains are relieved in shell-forgings by simply drilling out the nose.

We agree with Doctor Langenberg's view, that the improvement in physical properties is due to the release of internal strains. We believe that this release of strains is accompanied by a change in length of a longitudinal specimen, and if equipment of sufficient delicacy can be obtained the investigation will be continued along these lines.

Investigation along the lines suggested by Mr. Foley are being continued. Preliminary results show a decided improvement in physical properties of metal in blooms rested at 180° for several days. It is our intention to report to the Institute the results of this investigation as soon as it is completed.

When poor physical properties were unexpectedly encountered at the beginning of this work, we held an opinion similar to Mr. Comstock's; *viz.*, that these poor physical properties were due to strains set up by core drilling. To eliminate any strains produced by core drilling, specimens were cut from blooms by band saws, but equally poor properties followed which were improved by resting, as in the case of and to the same degree as drilled specimens. The question raised by Mr. Comstock relative to the resting of blooms before core drilling is answered in our remarks on Mr. Foley's discussion.

* Received Jan. 30, 1920.

Experimental Data Obtained on Charpy Impact Machine*

BY F. C. LANGENBERG,† WATERTOWN, MASS.

(Chicago Meeting, September, 1919)

IT is the purpose of this paper to present a limited amount of experimental data obtained on the Charpy impact machine. Several concrete examples will be given showing the relation existing between the results obtained upon the Charpy impact machine and the behavior of the parts tested when in service.

Although it is pointed out by G. Charpy and A. Cornu-Thénard[1] that it is first necessary to determine a satisfactory working process before studying the correlation between the results of these tests and the behavior of the parts tested in service, this ideal course cannot be strictly adhered to. If a certain part repeatedly fails in service and the available tests give no indication of its cause, advantage must be taken of any possible means that will solve the problem; therefore, even though the reliability of the results obtained by the Charpy impact machine have been questioned by many, it was necessary to use this instrument without giving careful consideration to its possible sources of error. It would seem that the conclusions arrived at by Charpy and Cornu-Thénard justify the assumption that the precision obtained by this machine is comparable to that obtained by the other more common forms of testing.

The second general conclusion arrived at by them, namely, that the impact test upon notching does not offer any direct correlation to the ordinary tests, is also borne out by many hundred determinations made in the laboratories at Watertown Arsenal. This statement, however, should be qualified as follows: Other ordinary tests include the static tensile test from bars taken in a longitudinal direction and the Brinell hardness. The author agrees that no law is yet known for the computation of the energy absorbed in the rupture of bars of the same material but of different dimensions. Although Charpy and Cornu-Thénard do not refer in their valuable article to the use of the Charpy machine in industrial practice, it should be stated that this instrument has been in constant use in the laboratory at Watertown Arsenal for several years, and is now employed for both routine and experimental testing.

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* Published by permission of the Ordnance Dept., U. S. A.

† Metallurgist, Watertown Arsenal.

[1] *Rev. de Mét.* (March–April, 1917).

Experience here has shown that the results obtained are very valuable for the production of material satisfactory for ordnance service. This statement is well supported by the fact that the authorities responsible for the satisfactory performance of ordnance material consider the results of sufficient importance to warrant the installation of a new Charpy machine of 300 kilogram-meter capacity which has been installed and is now in operation.

Prof. Henry M. Howe, in an article entitled "The Resilience Test,"[2] makes several statements to which exception must be taken. One of his questions "Can either the nicked or the unnicked test be made to give trustworthy results of fitness for nicked and unnicked impact service?" does not seem to be a proper statement of the problem because the difference between nicked and unnicked impact service cannot at present be defined. In other words, what definite change of section in a piece will be necessary in order that it will be subjected to nicked impact service? It would seem that this can be studied best by determining the influence of various nicks or notches on test pieces themselves. A great many results have been obtained on this subject, but they are far from being complete.

In order to answer his second question, "With materials of such a degree of heterogeneousness as ought reasonably to be provided for, can a single nicked test piece give a trustworthy measure of resistance?" it will first be necessary to define accurately the form and size of the test piece to be employed and the manner in which the stress is to be applied. A standard tensile specimen 0.505 in. (12.8 mm.) in diameter does not necessarily give the minimum physical properties of any element of the forging that it represents, and, in general, this is not what is desired. The safety of any piece in service depends on the total resistance to stress at the critical section, and it is the mean physical properties of a section equal to the critical section that are desired in the test to destruction of a representative specimen. A single tensile specimen does not by any means always give this information and it would not seem that a Charpy notched-bar transverse test should be expected to do so.

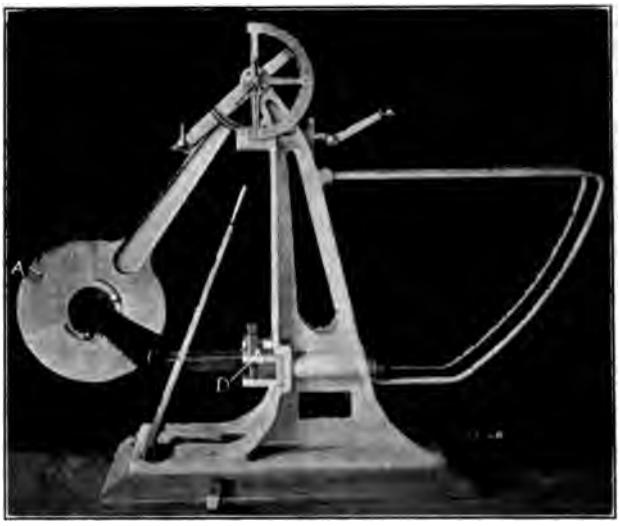
In regard to the third question asked by Professor Howe, "If not, is it more expedient to substitute a 2-in. threaded length of impact tensile test piece for the single nick, or to test of each material so large a number of test pieces as to insure detecting approximately the weakest section?" the last part of this query has been touched upon above. The threaded tensile specimen has been tried out in the laboratory at Watertown Arsenal and one illustration of the results obtained will be given in this paper. Experiments would indicate that a series of notches is far better than threading; as in the case of a threaded specimen the fracture

[2] *Met. & Chem. Eng.* (Sept. 15, 1917) 17.

must pass through the apex of some thread and as a result the area of fracture cannot be accurately determined. A series of notches accurately determine the area of metal that must be fractured before rupture occurs. The threaded specimen is used at the present time for certain forgings, which are themselves threaded in part or throughout.

Charpy Impact Machine and Test Bars

Although the construction and mode of operation of the Charpy machine is quite well known, the machine is shown in Fig. 1 for the benefit of those who have not seen it in operation. Several forms of Charpy test specimens are shown in Fig. 2.



B

Fig. 1.—Charpy impact machine.

The tension bars are fractured by screwing one end into the holder *A*, Fig. 1, which is a part of the pendulum itself. The lug *B* is then screwed on the other end of the specimen and when the pendulum is released the lug engages the arms *C* when the pendulum is in the vertical position. The pendulum continues its oscillation and the foot-pounds absorbed in breaking the specimen can readily be ascertained by noting the angle of deflection of the pendulum after breaking the specimen.

Tests Made on the Charpy Machine

In the notched-bar transverse test, the specimen is placed in the supports (*D*, Fig. 1), and the force applied by the knife of the pendulum, as indicated by the arrows in Fig. 2.

The notched-bar tensile test consists of an impact tensile test on a grooved tensile specimen, Fig. 2.

The Charpy tensile test consists of an impact tensile test on a uniform section tensile specimen. The diameter of this specimen in the stem is the same as the diameter of the Charpy notched bar at the base of the notch. In other words, to make the Charpy tensile specimen without the notch, the notched bar is turned down to a uniform diameter, which is that of the notched bar at the base of the notch.

Results of Tests on Chrome-nickel Steel Forging

The first tests to be described are those taken from a chrome-nickel steel forging approximately 5 in. (127 mm.) in diameter and 10 ft. (3 m.) long. This forging was threaded with a square thread throughout

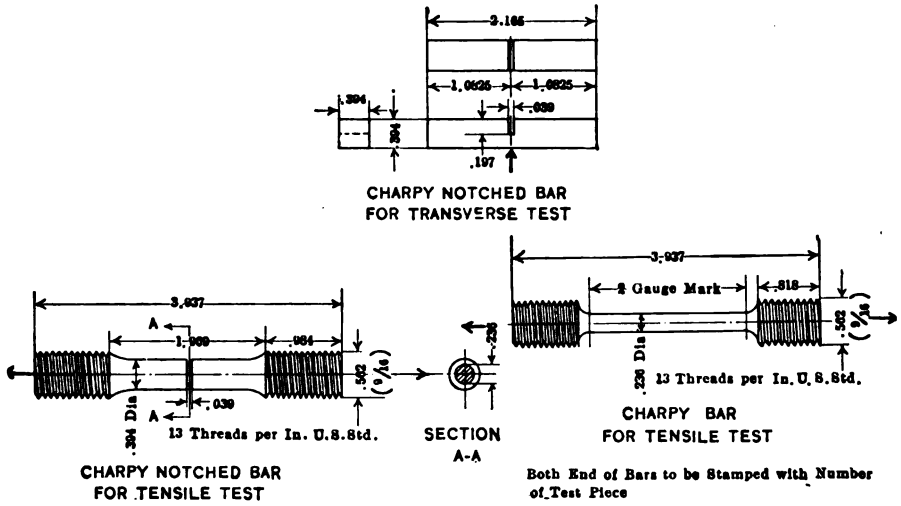


Fig. 2.—Forms of Charpy test specimens.

its length. The threads were approximately ¼ in. (6.35 mm.) deep. The composition of this forging was as follows: Carbon 0.43 per cent., manganese 0.64 per cent., silicon 0.19 per cent., sulfur 0.018 per cent., phosphorus 0.03 per cent., nickel 3.06 per cent., chromium 1.07 per cent.

This forging was purchased under the following specifications

Elastic limit	100,000 lb. per sq. in. (7030 kg. per sq. cm.).
Tensile strength	120,000 lb. per sq. in. (8436 kg. per sq. cm.).
Elongation	14.0 per cent.
Concentration of area	30.0 per cent.

and was accepted at the works of the manufacturer, after having passed

a satisfactory test, but failed in service. On its receipt in the laboratory, the following tests were carried out.

Static Tensile Test.—Tensile specimens tested in Emery hydraulic machine, shown in Fig. 3, were taken in the direction of the principal axis of the forging; or, in other words are longitudinal specimens. Tensile tests under a majority of present specifications would be taken in this manner.

FIG. 3.—RESULT OF STATIC TENSILE TESTS ON LONGITUDINAL SPECIMENS.

The results are as follows:

Specimen Marks	Elastic Limit, Lb. per Sq. In.	Yield Point, Lb. per Sq. In.	Tensile Strength, Lb. per Sq. In.	Elongation, Per Cent.	Contraction of Area, Per Cent.	Brinell Hardness Number
1	108,000	112,000	137,000	19.0	51.9	293
2	108,000	111,000	136,500	19.0	51.9	281
3	100,000	107,000	136,000	19.0	49.1	277
4	100,000	109,000	136,500	19.5	51.9	281
Mean	104,000	109,750	136,500	19.2	51.2	283

The specimens in Fig. 4, taken at right angles to principal axis of forging, or, in other words, transverse specimens, gave the following results:

Specimen Marks	Elastic Limit, Lb. per Sq. In.	Yield Point, Lb. per Sq. In.	Tensile Strength, Lb. per Sq. In.	Elongation, Per Cent.	Contraction of Area, Per Cent.	Brinell Hardness Number
1	.....	107,000*	116,500	2.5	5.7	285
2	85,000	99,000	121,500	3.5	5.7	277
3	85,000	95,000	120,000	4.0	5.7	277
Mean	85,000	97,000	119,350	3.3	5.7	280

* Extensometer readings not taken; result was not used in computing mean.

Charpy Notched-bar Transverse Test.—Fig. 5 shows the fractured specimens of the Charpy notched-bar transverse test. The lower row shows

FIG. 4.—RESULT OF STATIC TENSILE TESTS ON TRANSVERSE SPECIMENS.

those taken at right angles to the axis of the forging with the slot cut in the direction of forging; and the upper row, those taken in the direction

FIG. 5.—FRACTURED SPECIMENS OF NOTCHED-BAR TRANSVERSE TEST.

of the axis of the forging with the slot cut at right angles to the forging. The results are as follows:

Transverse Test		Longitudinal Test	
Number	Ft.-lb. Absorbed	Number	Ft.-lb. Absorbed
1	1.34	1	2.09
2	1.70	2	2.09
3	1.61	3	2.18
4	1.25	4	1.52
5	1.70	5	1.99
6	0.97	6	2.18
7	1.43	7	3.05
8	0.97	8	2.09
Mean	1.37	9	2.09
		Mean	2.14

FIG. 6.—NOTCHED AND UNNOTCHED BARS TAKEN ALONG AXIS OF FORGING.

Fig. 6 shows tensile Charpy specimens taken along the axis of the forging. Those in the upper row are not notched, while those in the lower row were notched with a diameter at the bottom of the notch equal to the diameter of the unnotched specimen. The results of the tests are as follows:

Specimens not Notched					Notched Specimens		
Number	Diameter, Inch	Ft.-lb. Absorbed	Elong. in 2 In., Per Cent.	Cont. of Area, Per Cent.	Number	Diameter, Inch	Ft.-lb. Absorbed
1	0.237	147.5	13.5	54.5	1	0.233	3.25
2	0.237	179.1	16.0	56.8	2	0.235	5.58
3	0.237	161.1	15.0	56.8	3	0.239	6.32
4	0.237	177.8	16.5	56.8	4	0.237	6.11
5	0.237	165.0	15.0	59.0	5	0.239	8.41
Mean	0.237	166.1	15.2	56.8	Mean	0.236	5.93

FIG. 7.—NOTCHED AND UNNOTCHED BARS TAKEN AT RIGHT ANGLES TO AXIS OF FORGING.

Fig. 7 shows tensile Charpy specimens taken at right angles to the axis of the forging. Those in the upper row are not notched while those in the lower row are notched with the diameter at the bottom of the notch equal to the diameter of the unnotched specimen. The results are as follows:

Specimens not Notched				Notched Specimens		
Number	Diameter, Inch	Ft.-lb. Absorbed	Elongation, Per Cent.	Number	Diameter, Inch	Ft.-lb. Absorbed
1	0.237	0.88*	0.9	1	0.238	1.43
2	0.237	5.69	1.0	2	0.237	2.47
3	0.237	3.75*	0.5	3	0.237	2.76
4	0.238	6.65	1.0	4	0.239	2.09
5	0.237	4.85	0.5	5	0.235	1.34
Mean		4.36	0.8	Mean		2.01

* Results not used in computing mean; fracture occurred at fillet making results comparable to those obtained on grooved specimens.

FIG. 8.—RESULTS OF TESTS OF MINIATURE FORGINGS.

Static Tests on Miniature Forgings.—In the static tests on miniature forgings, the specimens were pulled in the Emery hydraulic machine. They were machined from the forging as received from service and were made as nearly proportional as possible to the large forging. The two specimens on the left, Fig. 8, were taken in the direction of the axis of the forging, while the specimens on the right were taken at right angles to the axis of the forging. The results are as follows:

Longitudinal Miniature Forgings

Specimen	Elastic Limit, Lb. per Sq. In.	Tensile Strength, Lb. per Sq. In.	Elongation, Per Cent.	Contraction of Area, Per Cent.
1	117,750	146,650	5.5	31.1
2	120,000	151,100	8.0	24.4
Mean	118,875	148,875	6.7	27.7

Transverse Miniature Forgings

Specimen	Elastic Limit, Lb. per Sq. In.	Tensile Strength, Lb. per Sq. In.	Elongation, Per Cent.	Contraction of Area, Per Cent.
1	113,350	124,450	1.5	None
2	113,350	135,550	2.5	None
Mean	113,350	130,000	2.0	None

FIG. 9.—RESULTS OF IMPACT TENSILE TESTS ON MINIATURE CHARPY FORGINGS.

Impact Tensile Tests on Miniature Forgings.—In the Charpy tensile test on miniature forgings, both specimens were taken in the direction of the axis of forging; the results, shown in Fig. 9, are as follows:

FIG. 10.—MINIATURE FORGINGS BROKEN IN CHARPY MACHINE.

Specimen Number	Diameter, Inch	Ft.-lb. Absorbed	Elongation in 2 In., Per Cent.
1	0.240	23.15	2.5
2	0.240	17.28	1.5
Mean	0.240	20.21	2.0

Miniature Forging Broken in Charpy Machine.—The design of the thread was the same on the two specimens on the left, Fig. 10, as now used

on the forging The thread was changed on the two specimens on the right and rounded at the base. Fig. 11 illustrates the conditions present.

Description of Specimen	Diameter, Inch	Ft.-lb. Absorbed	Elongation in 2 In., Per Cent.	Mean Ft.-lb. Absorbed
Flat bottom thread.....	0.340	52.24	1.5	69.87
Flat bottom thread.....	0.340	87.50	3.0	
Round bottom thread.....	0.340	38.55	1.0	44.59
Round bottom thread.....	0.340	50.63	1.5	

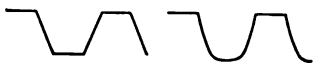


FIG. 11.—FORMS OF SCREW THREADS.

In Fig. 12, from left to right, the first specimen shows the fractured end of static tensile specimen taken in longitudinal direction. The test indicates good ductility, elastic limit, and tensile strength. The second specimen represents a static tensile specimen taken at right angles to the first, or in other words, a transverse specimen. This bar had no ductility but good elastic limit, and tensile strength.

The third specimen represents an unnotched longitudinal Charpy specimen; it had high shock strength and good ductility. The fourth

FIG. 12.—TYPE BARS FROM DEFECTIVE FORGING.

specimen represents a notched longitudinal Charpy specimen, which had low shock strength. The fifth specimen represents a miniature forging taken in a longitudinal direction and broken as a tensile specimen in the Charpy machine; it had low shock strength.

An examination of the physical tests in Tables 1 and 2 enables the statement of several facts.

1. The longitudinal static tensile specimens show good physical qualities and meet the required specifications in a satisfactory manner. The yield point and elastic limit are only separated by 5750 lb. (2608 kg.).

TABLE 1.—*A Summary of Results on the Forging as Received from Service*

Nature of Tests	Description of Specimen	Yield Point, Lb. per Sq. In.	Elastic Limit, Lb. per Sq. In.	Tensile Strength, Lb. per Sq. In.	Elongation, Per Cent.	Contraction of Area, Per Cent.	Ft.-lb. Absorbed
Static tensile, mean of four tests.	0.505 in. standard tensile, longitudinal.	109,750	104,000	136,500	19.2	51.2	
Static tensile, mean of three tests.	0.505 in. standard tensile, transverse.	97,000	85,000	119,350	3.3	5.7	
Charpy notched bar, transverse test, mean of eight tests.	Notched bar, longitudinal.						2.14
Charpy notched bar, transverse test, mean of eight tests.	Notched bar, transverse.						1.37
Charpy tensile test, mean of five tests.	Small tensile bar without notch, 0.237 in. diameter, longitudinal.				15.2	56.8	166.1
Charpy notched bar, tensile test, mean of five tests.	Small tensile bar with notch 0.237 in. at bottom, longitudinal.						5.93
Charpy tensile test, mean of five tests.	Small tensile without notch, 0.237 in. diameter, transverse.				0.8		4.36
Charpy notched bar, tensile test, mean of five tests.	Small tensile bar with notch, 0.237 in. diameter, transverse.						2.01

TABLE 2.—*Miniature Forging*

Nature of Test	Yield Point, Lb. per Sq. In.	Tensile Strength, Lb. per Sq. In.	Elongation, Per Cent.	Contraction of Area, Per Cent.	Ft.-lb. Absorbed
Static longitudinal, mean of two tests.....	118,875	148,875	6.7	27.7	
Static transverse, mean of two tests.....	113,350	130,000	2.0	None	
Charpy tensile tests, longitudinal, mean of two tests.					20.21
Charpy tensile test, longitudinal, diameter at bottom of thread 0.340 in., flat bottom thread.....					69.87
Charpy tensile test, longitudinal, diameter at bottom of thread 0.340 in., round bottom thread....					44.59

2. The transverse static tensile specimens show a lower elastic limit and ultimate strength than the longitudinal tensile specimens, but are particularly marked by the absence of ductility. The yield point and tensile strength differ by 12,000 lb. (5443 kg.).

3. The Charpy strength, as determined by the notched-bar transverse test, is very low, being comparable to a good cast iron. The specimens taken from the forging in a longitudinal direction gave approximately 60 per cent. greater resistance than those taken in transverse direction. Similar forgings recently manufactured have a Charpy strength of 24 to 28 ft.-lb.; or using this test as a criterion, they have over ten times as great shock-resisting power. No trouble to date has been experienced with this material and the manufacture of a satisfactory forging for the particular service required was made possible by the use of the Charpy machine.

4. The small tensile unnotched Charpy bars taken longitudinally absorbed 166.1 ft.-lb.; the diameter of these specimens was 0.237 in. The small tensile bars with a notch, the diameter at the base of the notch being 0.237 in., absorbed 5.93 ft.-lb. The very apparent fact is that the notch seriously reduced the shock strength, the notched bars offering one-twenty-eighth of the resistance of the unnotched specimens. Neglecting the merits and demerits of the respective specimens, it can be stated positively that abrupt change of section acts in some very pronounced manner to reduce the resistance to shock.

5. The small unnotched Charpy tensile specimens taken in a transverse direction gave a mean value of 4.36 ft.-lb. absorbed, whereas the notched tensile specimens taken in the same direction gave 2.01 ft.-lb. absorbed, or a reduction of approximately one-half. The difference between the notched and unnotched specimens was much less when the specimens were taken in the transverse direction. It is to be noted in this connection that the ductility was good in the longitudinal direction, but practically zero in the transverse direction. It would seem that a very close relation exists between ductility and the effect of abrupt section changes.

6. It is also of interest to note that the longitudinal specimens notched gave a mean value of 5.93 ft.-lb. absorbed, whereas the unnotched transverse specimens gave 4.36 ft.-lb. absorbed.

7. As the influence of the notch on specimens subjected to shock was so apparent, a number of miniature forgings were made, as previously described. These miniature forgings had dimensions proportional to the service forging. Static tensile tests on these forgings showed that the yield and tensile strength were increased by the presence of the threads, whereas the ductility was reduced. This is true for both longitudinal and transverse specimens, and is in agreement with our present knowledge that a notch or thread will increase the yield point and tensile strength on the reduced section caused by notching or threading.

8. Miniature forgings were then taken in a longitudinal direction to approximate service conditions. The foot-pounds absorbed in the rupture of these specimens gave a mean value of 20.21: Unnotched bar, 0.237 in. in diameter, 166.1 ft.-lb. absorbed; notched bar 0.237 in. at base of notch, 5.93 ft.-lb absorbed; miniature threaded forging 0.237 in. at root of thread, 20.21 ft.-lb absorbed. The notch in the notched-bar specimen was continuous, whereas in the threaded specimen the fracture must pass through a thread or over a greater area than is the case with the regular notched Charpy tensile specimen. These results show beyond doubt that threading has served to bring into play the factors associated

FIG. 13.—MICROGRAPH OF FORGING. × 335.

with abrupt change of section, and it is reasonable to assume that these factors are active under service conditions, as the fractures of the forgings that failed in service were very similar to the fracture of the miniature forging broken in the Charpy machine. Furthermore, no ductility was displayed by either, after rupture, which shows conclusively that the properties exhibited by the longitudinal static tensile specimen were not utilized previous to failure.

9. The ordinary tensile tests showed a complete lack of ductility at right angles to the axis of the forging. Our present knowledge would, therefore, lead us to conclude that brittleness would be experienced at right angles to the principal axis. The unnotched Charpy tensile specimens taken at right angles to the axis of the forging verify this conclusion.

The apparent effect of notching a bar taken in the longitudinal direction is to weaken it under shock. It is assumed that the effect of notching is to seriously augment the stresses at right angles to the main axis of the forging, and as brittleness is present in this direction, failure occurs at unexpectedly low stresses.

Fig. 13 shows, magnified 335 diameters, the general structure of the forging as taken from service. The original martensitic formation caused by quenching still persists, which is in itself an indication of brittleness, and the presence of such a structure should have been sufficient grounds for recommending retreatment.

Results of Tests of Second Forging

The following results were obtained from a forging that was a duplicate of the first, and which also failed in service, and an inspection of these figures, together with a comparison with the results on forging No. 1, may be of interest.

Table 3.—*Static Tensile Test*

Specimen Mark	Elastic Limit, Lb. per Sq. In.	Yield Point, Lb. per Sq. In.	Tensile Strength, Lb. per Sq. In.	Elongation, Per Cent.	Contraction of Area, Per Cent.	Brinell Hardness Number
Longitudinal Specimen						
1	110,000	114,000	140,000	16.5	46.2	285
2	105,000	114,000	140,500	16.0	46.2	285
3	100,000	111,000	139,000	17.5	46.2	285
4	104,000	109,000	138,000	17.5	46.2	285
Mean	104,750	112,000	139,250	16.9	46.2	285
Transverse Specimen						
1	92,000	99,000	108,542	1.5	5.5	285
2	92,000	100,000	110,000	1.5	5.7	262
3	95,000	100,000	114,500	1.5	5.7	285
Mean	93,000	99,666	111,014	1.5	5.6	277

Table 4.—*Charpy Notched-bar Transverse Test*

Specimens Taken Longitudinally, Ft.-lb. Absorbed	Specimens Taken Transversely, Ft.-lb. Absorbed	Specimens Taken Longitudinally, Ft.-lb. Absorbed	Specimens Taken Transversely, Ft.-lb. Absorbed
2.09	0.70	1.61	1.34
1.61	1.99	1.99	1.16
1.99	1.25	2.09	1.06
2.09	0.79	Mean, 1.94	Mean, 1.18
2.09	1.16		

8. Miniature forgings were then taken in a longitudinal direction to approximate service conditions. The foot-pounds absorbed in the rupture of these specimens gave a mean value of 20.21: Unnotched bar, 0.237 in. in diameter, 166.1 ft.-lb. absorbed; notched bar 0.237 in. at base of notch, 5.93 ft.-lb absorbed; miniature threaded forging 0.237 in. at root of thread, 20.21 ft.-lb absorbed. The notch in the notched-bar specimen was continuous, whereas in the threaded specimen the fracture must pass through a thread or over a greater area than is the case with the regular notched Charpy tensile specimen. These results show beyond doubt that threading has served to bring into play the factors associated

FIG. 13.—MICROGRAPH OF FORGING. × 335.

with abrupt change of section, and it is reasonable to assume that these factors are active under service conditions, as the fractures of the forgings that failed in service were very similar to the fracture of the miniature forging broken in the Charpy machine. Furthermore, no ductility was displayed by either, after rupture, which shows conclusively that the properties exhibited by the longitudinal static tensile specimen were not utilized previous to failure.

9. The ordinary tensile tests showed a complete lack of ductility at right angles to the axis of the forging. Our present knowledge would, therefore, lead us to conclude that brittleness would be experienced at right angles to the principal axis. The unnotched Charpy tensile specimens taken at right angles to the axis of the forging verify this conclusion.

2. The Charpy notched-bar transverse tensile test gave almost identical results on the two forgings.

3. The results of the Charpy notched-bar tensile test were nearly alike for the specimens taken in a longitudinal direction; for those taken in the transverse direction, the specimens from the second forging gave inferior results, approximating 50 per cent. of the strength shown by the specimens from the first.

4. The results of the Charpy tensile specimens without notches agreed very closely with results obtained on first forging for both those taken in

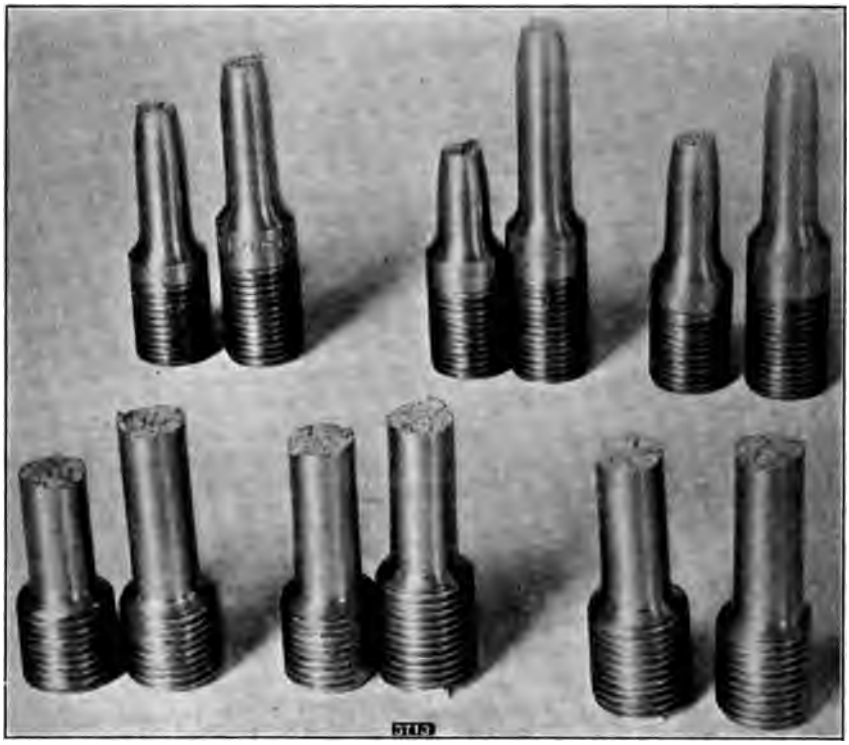


FIG. 14.—APPEARANCE OF HEAT-TREATED BARS AFTER FRACTURE; TOP ROW ARE LONGITUDINAL SPECIMENS, BOTTOM ROW ARE TRANSVERSE.

the longitudinal and the transverse direction. The Charpy notched-bar tensile tests for the second forging, taken in the longitudinal direction, gave a mean of 5.59 ft.-lb. absorbed, whereas the Charpy tensile specimens without a notch taken in the transverse direction gave 7.22 ft.-lb. absorbed. Here again is evidence that notching the longitudinal test specimen brings into play the properties exhibited by the transverse unnotched specimen.

Results of Tests Made after Heat-treating Forging

As the results of the tests on the two forgings were in such good agreement, and as they apparently suffered from the same defects, the first was selected for further work. This additional investigation consisted in heat-treating a section of the forging and determining its physical properties by a series of tests similar to those already described. The heat treatment consisted in heating one of the broken sections of the

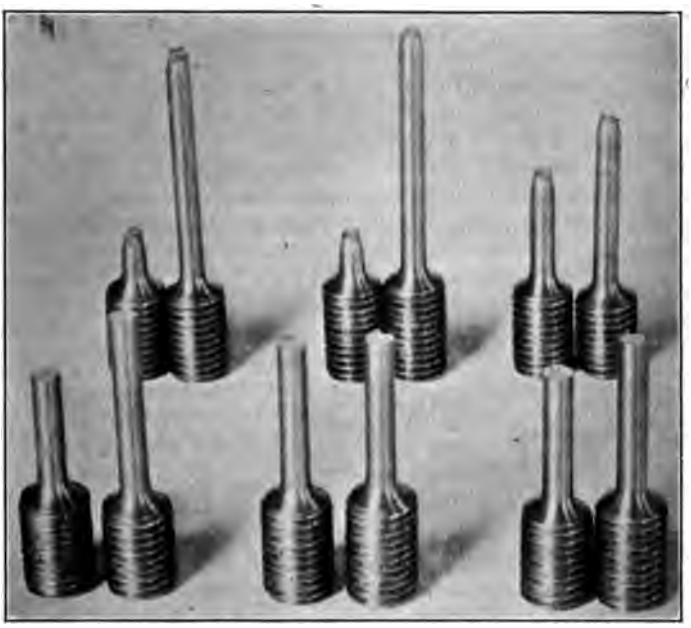


Fig. 15.—Appearance of heat-treated bars after Charpy tensile test.

forging about 2 ft. in length to 800° C., followed by quenching in water. After quenching in water, the piece was tempered 4 hr. at 575° C. One-half of this section was cut up for test, and the remaining piece tempered two additional hours at 650° C.

Static Tensile Test.—Fig. 14 illustrates the appearance of the specimens after fracture, the top row showing the longitudinal specimens and the bottom row the transverse.

Charpy Tensile Test.—Fig. 15 illustrates the appearance of the specimens after fracture, the top row showing the longitudinal tests and the bottom row showing the transverse.

The microstructure of the material of the forging after quenching and drawing at 575° C. showed a very fine structure consisting of sorbite and

TABLE 8.—*Static Tensile Tests on Heat-treated Bars*

Specimen	Elastic Limit, Lb. per Sq. In.	Yield Point, Lb. per Sq. In.	Tensile Strength, Lb. per Sq. In.	Elongation, Per Cent.	Contraction of Area, Per Cent.	Brinell Hardness Number
LONGITUDINAL SPECIMENS						
1	105,000	110,000	135,500	19.5	49.1	285
2	105,000	110,000	136,500	17.0	49.1	285
3	105,000	109,000	135,500	18.5	46.2	269
Mean	105,000	109,666	135,833	18.2	48.1	279
TRANSVERSE SPECIMENS						
1	100,000	105,000		3.5		277
2	100,000	104,000	129,500	6.5	13.3	277
3	100,000	105,000	131,000	9.0	9.5	269
Mean	100,000	104,666	130,250	6.3	11.4	274

TABLE 9.—*Charpy Tensile Tests on Heat-treated Bars*

Specimen	Longitudinal Specimens			Transverse Specimens		
	Ft.-lb. Absorbed	Elongation, Per Cent.	Contraction of Area, Per Cent.	Ft.-lb. Absorbed	Elongation, Per Cent.	Contraction of Area, Per Cent.
1	176.3	16.0	50.0	65.60	7.0	9.0
2	161.9	14.5	54.5	116.80	10.5	18.1
3	184.7	16.5	56.8	50.63	5.0	6.8
Mean	174.3	15.6	53.7	77.67	7.5	11.3

TABLE 10.—*Charpy Notched-bar Transverse Test*

Specimen	Specimen Taken in Longitudinal Direction, Ft.-lb. Absorbed	Specimen Taken in Transverse Direction, Ft.-lb. Absorbed
1	3.25	2.09
2	2.18	1.99
3	2.56	3.05
4	3.25	2.09
Mean	2.81	2.30

TABLE 11.—*Charpy Notched-bar Tensile Test*

Specimen	Specimen Taken in Longitudinal Direction, Ft.-lb. Absorbed	Specimen Taken in Transverse Direction, Ft.-lb. Absorbed
1	14.08	3.15
2	10.72	2.09
3	12.86	7.74
Mean	12.55	4.32

troostite. The faint outlines of large grains were visible. This network could not have resulted from the heating for quenching, as the temperature was too low. It was probably the result of some earlier treatment, the results of which were not obliterated entirely by the subsequent heating above the critical range.

TABLE 12.—*Summary of Tests on Heat-treated Forging*

Nature of Test	Description of Specimen	Elastic Limit, Lb. per Sq. In.	Yield Point, Lb. per Sq. In.	Tensile Strength, Lb. per Sq. In.	Elongation, Per Cent.	Contraction of Area, Per Cent.	Foot-pounds Absorbed
Static tensile, mean of three tests.	0.505 in. diameter, longitudinal.	105,000	109,666	135,833	18.2	48.1	
Static tensile, mean of three tests.	0.505 in. diameter, transverse.	100,000	104,666	130,250	6.3	7.6	
Charpy notched-bar transverse test, mean of four tests.	Bars taken in longitudinal direction.						2.81
Charpy notched-bar transverse test, mean of four tests.	Bars taken in transverse direction.						2.30
Charpy tensile test, mean of three tests.	0.237 in. diameter, longitudinal.				15.6	53.7	174.3
Charpy tensile test, mean of three tests.	0.237 in. diameter, transverse.				7.5	11.3	77.67
Charpy notched-bar tensile test, mean of three tests.	0.237 in. diameter at bottom of notch, longitudinal.						12.55
Charpy notched-bar tensile, mean of three tests.	0.237 in. diameter at bottom of notch, transverse.						4.32

TABLE 13.—*Results of Static Tensile Tests after Second Heat Treatment*

LONGITUDINAL SPECIMENS

Specimen Mark	Elastic Limit, Lb. per Sq. In.	Yield Point, Lb. per Sq. In.	Tensile Strength, Lb. per Sq. In.	Elongation, Per Cent.	Contraction of Area, Per Cent.	Brinell Hardness Number
1	94,000	98,000	122,000	20.0	54.6	255
2	93,000	96,000	120,500	21.0	54.6	248
3	92,000	95,000	120,000	21.0	54.6	248
Mean	93,000	96,333	120,833	20.6	54.6	250
TRANSVERSE SPECIMENS						
1	90,000	93,000	113,500	16.5	34.0	241
2	93,000	95,000	119,500	16.0	34.0	241
3	93,000	96,000	120,500	14.5	34.0	248
Mean	92,000	94,666	119,500	15.6	34.0	243

Results of Tests after Second Heat Treatment

The forging was heated to 800° C., quenched in water, drawn 4 hr. at 575° C., and two additional hours at 650° C.

Static Tensile Test.—Fig. 16, in the top row, shows specimens taken in the longitudinal direction and in the bottom row shows the transverse specimens, which were submitted to the static tensile tests.

Fig. 16.—Fractures of specimens after second heat treatment submitted to static tensile test.

Table 14.—*Results of Charpy Tensile Test after Second Heat Treatment*

Specimen	Longitudinal Specimens			Transverse Specimens		
	Ft.-lb. Absorbed	Elongation, Per Cent.	Contraction of Area, Per Cent.	Ft.-lb. Absorbed	Elongation, Per Cent.	Contraction of Area, Per Cent.
1	182.9	18.0	56.8	163.2	16.0	40.9
2	177.2	18.0	56.8	176.0	17.5	43.1
3	165.5	16.5	59.0	168.1	16.0	40.9
Mean	175.2	17.5	57.5	169.1	16.5	41.6

TABLE 15.—*Charpy Notched-bar Transverse Test after Second Heat Treatment*

Specimen	Specimen Taken in Longitudinal Direction, Ft.-lb. Absorbed	Specimen Taken in Transverse Direction, Ft.-lb. Absorbed
1	11.19	2.27
2	9.90	5.06
3	11.07	5.79
4	12.02	5.16
Mean	11.04	4.57

Charpy Tensile Test.—Fig. 17, in the top row, shows the longitudinal specimens and in the bottom row shows transverse specimens that were submitted to the Charpy tensile test.

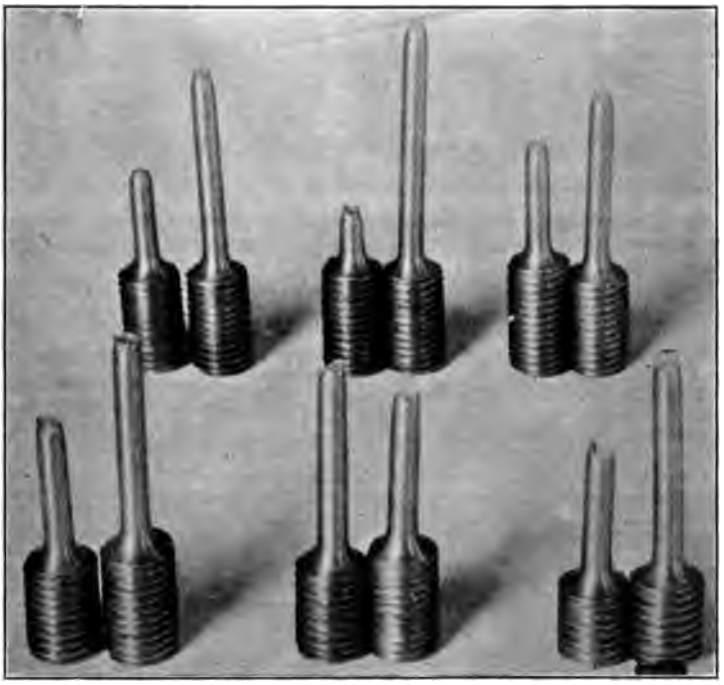


FIG. 17.—FRACTURES OF SECOND HEAT-TREATED SPECIMENS SUBMITTED TO CHARPY TENSILE TEST.

TABLE 16.—*Charpy Tensile Notched-bar Tests after Second Heat Treatment*

Specimen	Specimen Taken in Longitudinal Direction, Ft.-lb. Absorbed	Specimen Taken in Transverse Direction, Ft.-lb. Absorbed
1	14.58	8.76
2	18.60	6.65
3	11.54	7.52
Mean	14.91	7.64

FIG. 18.—FRACTURES OF TRANSVERSE BARS SUBMITTED TO STATIC TENSILE TESTS. THE UPPER ROW SHOWS THOSE DRAWN AT 650° C.; THE LOWER ROW THOSE DRAWN AT 575° C.

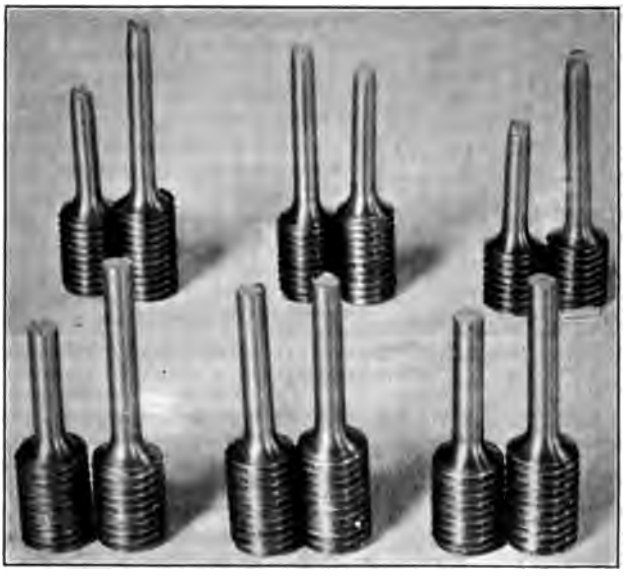


FIG. 19.—FRACTURES OF TRANSVERSE BARS SUBMITTED TO CHARPY TENSILE TESTS; BARS IN BOTTOM ROW TEMPERED AT 575° C., BARS IN TOP ROW TEMPERED AT 650° C.

TABLE 17.—*Mean Results on Static Tensile Tests Taken in Transverse Direction*

Specimen	Elastic Limit, Lb. per Sq. In.	Yield Point, Lb. per Sq. In.	Tensile Strength, Lb. per Sq. In.	Elongation, Per Cent.	Contraction of Area, Per Cent.	Brinell Hardness Number
Drawn at 575° C.....	100,000	104,666	130,250	6.3	7.6	274
Drawn at 650° C.....	93,000	94,666	119,500	15.6	34.0	243

Fig. 19 groups the fractured Charpy tensile tests taken in a transverse direction from the forging in question after quenching at 800° C. in water and tempering at 575° and 650° C. respectively. The bottom row was taken from material tempered at 575° C. and the top row from material tempered at 650° C.

TABLE 18.—*Mean Results of Charpy Tensile Tests on Transverse Bars*

Specimen	Ft.-lb. Absorbed	Elongation, Per Cent.	Contraction of Area, Per Cent.
Drawn at 575° C.....	77.67	7.5	11.3
Drawn at 650° C.....	169.1	16.5	41.6

TABLE 19.—*Summary of Tests on Second Heat-treated Forgings*

Nature of Test	Description of Specimen	Elastic Limit, Lb. per Sq. In.	Yield Point, Lb. per Sq. In.	Tensile Strength, Lb. per Sq. In.	Elongation, Per Cent.	Contraction of Area, Per Cent.	Foot-Pounds Absorbed
Static tensile, mean of three tests.	0.505 in. diam., longitudinal.	93,000	96,333	120,833	20.6	54.6	
Static tensile, mean of three tests.	0.505 in. diam., transverse.	92,000	94,666	119,500	15.6	34.0	
Charpy notched-bar transverse test, mean of four tests.	Bars taken in longitudinal direction.						11.04
Charpy notched-bar transverse test, mean of four tests.	Bars taken in transverse direction.						4.57
Charpy tensile test, mean of three tests.	0.237 in. diam., longitudinal.				17.5	57.5	175.2
Charpy tensile test, mean of three tests.	0.237 in. diam., transverse.				16.5	41.6	169.1
Charpy notched-bar tensile test, mean of three tests.	0.237 in. diam. at bottom of notch, longitudinal.						14.91
Charpy notched-bar tensile test, mean of three tests.	0.237 in. diameter, at bottom of notch, transverse.						7.64

The microstructure of the forging after the 650° C. tempering operation was distinctly sorbitic. A comparison of this microstructure with that of the forging as received showed very clearly that decided improvement was brought about by the higher drawing temperature.

Results and Values of Tests

In Fig. 20, the results of the tests taken in the longitudinal direction are shown graphically; values obtained for a particular test are represented by ordinates. Three series of tests are shown; longitudinal tests on the forging as received from service marked ⊙; longitudinal tests

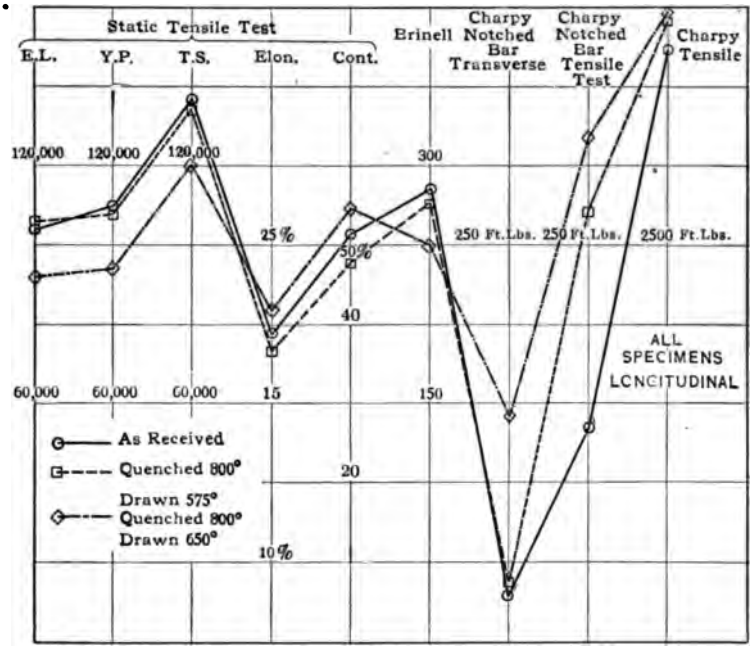


Fig. 20.—Results of tests of longitudinal specimens.

on the forging after treatment consisting in water quenching from 800° C., followed by tempering at 575° C., for 4 hr. marked ⊡; and longitudinal tests on the forging after treatment consisting of water quenching from 800° C. followed by tempering at 575° C., for 4 hr. and 2 hr. at 650° C., marked ◇. Under present methods of purchasing material, only the first five determinations are generally specified; namely, elastic limit, yield point, tensile strength, elongation, and contraction. Experience indicates that a low figure for the Charpy notched-bar transverse test indicates brittleness under suddenly applied stresses. The problem thus resolves itself into one of raising the "Charpy strength" by controlling one or all of the first five qualities. It is evident from the graph that

lowering the elastic limit by some 10,000 lb. (4535 kg.) by proper tempering after quenching has been associated with a great increase in the "Charpy strength."

To summarize, the material with a lower elastic limit, lower tensile strength, slightly higher ductility, and lower Brinell hardness has a much higher shock strength, as determined by the Charpy notched-bar specimen, either tensile or transverse. These slight variations in the ordinary static tensile properties, as determined on longitudinal specimens, do not properly account for the great differences present in shock strength, as determined by the notched specimens. The above statement has

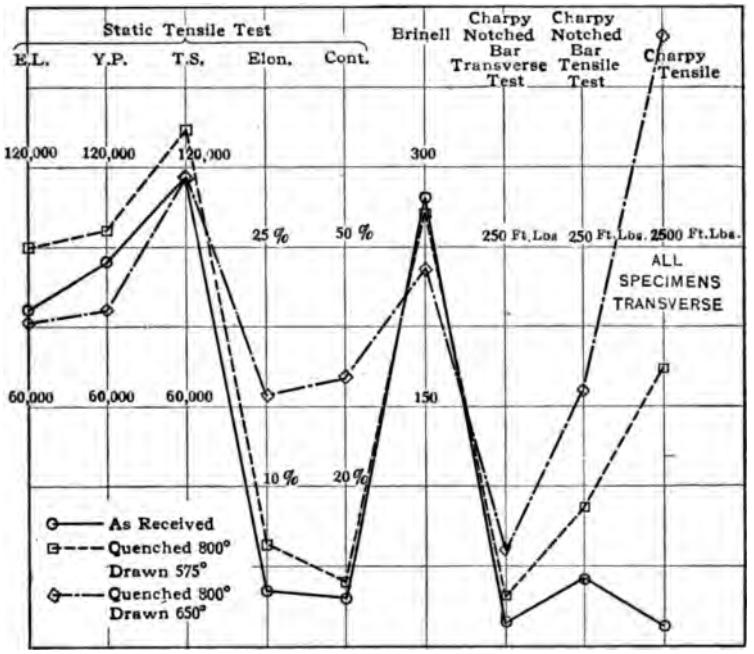


FIG. 21.—RESULTS OF TESTS OF TRANSVERSE SPECIMENS.

been verified by many observations, and it can be very definitely stated that no definite relation can be observed between the results of the Charpy test on notched bars and the results obtained in the static tensile tests on bars taken in a longitudinal direction.

Attention is invited to a similar graph shown in Fig. 21. The results are shown of a similar series of tests in which all specimens were taken at right angles to the principal axis of the forging; or in other words, were transverse specimens. An inspection of this graph will show that the only consistent relation between any of the Charpy tests and ordinary static properties lies between ductility and the shock strength. Again referring to the tests taken in a longitudinal direction, a lower elastic

limit or yield point seemed to bring about a higher shock resistance. It seems reasonable to presume that this relation is indirect only. A lower elastic limit, as determined in a static tensile specimen, if brought about by longer or more complete tempering, will result also in more ductility in the transverse direction. It is the ductility in the transverse direction that is intimately associated with the shock strength in the longitudinal direction if any change of section is encountered, which was the case in the forgings under discussion.

As pointed out in a previous part of this paper, it is very difficult to distinguish between nicked and unnicked service. As an illustration of this point, a number of tests are shown which were taken from a forging that failed in service when subjected to only a very low stress. The forging in question did not have any abrupt section changes similar to those in the forgings previously described. All changes of section were well relieved by fillets and no sharp notches or nicks were present Four static tensile specimens were taken, three in the transverse direction, marked respectively 1, 2, and 3, and one in the longitudinal direction marked 66. The results are as tabulated below. The Brinell hardness was taken on the end of the broken tensile specimens.

Marks	Elastic Limit, Lb. per Sq. In.	Tensile Strength, Lb. per Sq. In.	Elongation in 2 In., Per Cent.	Contraction of Area, Per Cent.	Brinell Hardness Number
1	45,000	62,500	1.5	1.8	179
2	45,000	69,500	1.5	1.8	187
3	42,000	61,500	1.0	1.8	183
66	60,000	93,333	25.0	65.0	179

It is to be noted that for the transverse tensile specimens the elastic limit is 45,000 lb. per sq. in (3163 kg. per sq. cm.) for specimens 1 and 2; 42,000 lb. (2952 kg.) for specimen 3; and for the specimen taken in the longitudinal direction, the elastic limit was 60,000 lb. per sq. in. (4218 kg. per sq. cm.). It is to be further noted that great discrepancies exist between the tensile strength of the tensile specimens taken in the transverse direction and the tensile specimens taken in the longitudinal direction. Furthermore, the tensile specimens taken in the transverse direction show practically no ductility, while the longitudinal specimen showed very good elongation and contraction, and passed all the requirements under which the material was purchased; namely, elastic limit, 53,000 lb. per. sq. in. (3726 kg. per sq. cm.); tensile strength, 93,000 lb. per sq. in. (6538 kg. per sq. cm.); 18 per cent. elongation after rupture, 30 per cent. contraction of area. Here again is a definite illustration of the fact that our present system of testing and the manner of taking tests are not adequate to insure uniformly satisfactory material when the service stresses do not occur in a longitudinal direction.

Four notched-bar transverse Charpy tests were taken from the forging. These specimens were cut in the longitudinal direction and gave the following results: 1.70, 1.25, 1.16, 1.25 ft.-lb. absorbed; mean, 1.34 ft.-lb.

The Charpy qualities of this material are approximately equal to those obtained on a good grade of gray iron. If tensile specimens taken in the longitudinal direction only had been tested, no apparent relation would have existed between the very low Charpy strength and the static properties, namely, elastic limit, tensile strength, elongation, and contraction. The static properties obtained from the specimens taken in the transverse direction are sufficient in themselves to warrant the rejection of the piece.

As all of the forgings discussed have suffered from very low ductility when static tensile tests were taken in the transverse direction, while at the same time the Charpy test has been low, it will doubtless occur to many that the low Charpy test is an indication of poor ductility in some plane of the forging. This assumption is further supported by the fact that a great many other forgings examined have given results very similar to those described in this paper. A more detailed discussion of this point, together with some data on the rate of stress application and its influence on notched specimens, will be reserved for another discussion on this subject.

Supplementary Note

Attention is invited to the fact that in this report the results of the Charpy impact test are recorded as foot-pounds absorbed. This means the actual number of foot-pounds absorbed in breaking the specimen. For convenience, the results are generally reported as foot-pounds absorbed per square inch. This value can be obtained by noting the area of the specimen broken and the foot-pounds required to break this area. For the notched transverse Charpy bar, multiplying the results recorded by 12.7 will give with reasonable accuracy the computed value of foot-pounds absorbed per square inch.

It has been proved that the same laws of similarity that hold for tensile specimens do not apply for Charpy specimens. In other words, if a small transverse Charpy bar having the dimensions shown in this report absorbs 2 ft.-lb. of energy in rupture, it does not follow that a Charpy bar having similar dimensions to the small bar but having 1 sq. in. of material below the notch will absorb 12.7 times 2 ft.-lb. In other words, data are not yet available that will allow the computing of the force of rupture under shock of a large specimen from the data obtained upon a small specimen. The results of the Charpy test must, at the present time, be considered comparative; but the fact that they must be considered comparative only, at the present time, does not in any way detract from their value in the study of the properties of any material.

Before concluding, the author of the above note wishes to state that the above investigation was inaugurated by General C. B. Wheeler, who was one of the first to realize the importance of the shock test in the proper selection of ordnance material.

It was through the efforts of General Wheeler that the first Charpy machine was installed in the testing laboratory at Watertown Arsenal, and it is believed that this machine was one of the first, if not the first, installed in this country.

Brig. Gen. T. C. Dickson and Col. C. M. Wesson have closely followed the results obtained upon the small Charpy impact machine, and have followed closely its use in the development of satisfactory ordnance materials. It was through their efforts that the present 300-kgm. Charpy machine was procured and installed.

The present commanding officer of this Arsenal, Brig. Gen. T. C. Dickson, has been instrumental in the development of a program of tests to be carried out on the large Charpy machine, and it is hoped that these results, together with other results obtained since the writing of the above note, can be presented at some later date.

The author wishes to express his appreciation for the assistance and help received from the above named officers, as well as from many other officers in the Ordnance Department whose names are not mentioned. Their assistance has not been confined alone to the securing of the necessary apparatus, but has consisted as well of many valuable suggestions and criticisms. The author wishes to mention three members of the laboratory personnel without whose help the obtaining of the results would have been impossible; namely, Mr. J. A. Powell, Mr. Robert Ramsey and Miss M. H. Neagle.

DISCUSSION

Geo. F. Comstock,* Niagara Falls, N. Y. (written discussion†).—In this exposition of experimental data, the author infers, indirectly at least, that by the Charpy test information was obtained that led to the improvement of certain ordnance parts and that this improvement could not have been attained without the Charpy test. On examination of the data, however, it appears that the important point of the whole matter was that while the material had been accepted from the results of longitudinal tensile tests, it was really transverse ductility that was desired. The writer has no desire to attempt to detract from the value of the paper or to discredit the Charpy test in any way, but he would like to ask the author if he will not give more clearly his reasons for preferring the Charpy impact test for investigational and control work of this kind, over

* Metallurgical Engineer, Titanium Alloy Mfg. Co. † Received Sept. 29, 1919.

the more usual tensile test made in a transverse direction. The tensile test pieces are more cheaply made than the square notched impact bars; and from the former the limit of elasticity and the ultimate strength as well as the ductility are obtained, while in the Charpy test only one property is measured. It does not seem that the author has shown a sufficient reason for favoring the latter test, when both it and the transverse tensile test served his purpose equally well in indicating the proper metal and treatment required for his special forging, and it is hoped that this reason will be given more clearly as a matter of useful general information.

A. G. ZIMERMANN,* Washington, D. C. (written discussion†).—The Watertown Arsenal is to be congratulated on performing some excellent pioneer work in accumulating data relative to the shock resistance of material used for ordnance purposes. Doctor Langenberg's work serves to demonstrate the correctness of two principles which should be the basis of all inspection by the consumer of materials subject to shock where impact testing machines are not available or where the data on results with impact testing machines are not considered conclusive enough to form the basis for specifications.

These two principles are: (*a*) Acceptance of forgings need not necessarily be based on the performance of test bars pulled in the direction of the stresses that will be applied in service. (*b*) Material that has normal ductility in its weakest plane is a good shock-resisting material.

The first principle follows the fundamental idea that if the most unfavorable test of the material is definitely known, the inspector has good grounds for assuming that the remainder of the forging is as good or better, when viewed from any other angle, as the results on which he accepted the forging. This is the more reasonable since, as pointed out by Doctor Langenberg, it is not always safe to say that no transverse stresses will be set up when, theoretically, the stresses of service are longitudinal.

If the location of test specimens is to be determined beforehand, irrespective of the method of forging, there would be an undue incentive to the manufacturer to do excessive forging in the direction of the test specimen. It is undoubtedly true that the more forging material receives in one direction, beyond a certain point, the less will be its ductility at right angles to this direction—and this effect is much more marked in basic steel than in acid steel. Since sudden shock appears to search out the weakest plane of a forging, it is apparent that normal ductility in one direction should not be sacrificed for a good showing in ductility in another direction.

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* Lieutenant Commander, U. S. N., U. S. Naval Gun Factory.

† Received Oct. 6, 1919.

Doctor Langenberg's results, showing that the impact test follows somewhat the course indicated by the results of the transverse static test, is borne out by the fact that the Naval Bureau of Ordnance has been following the above principles of inspection for some time and a consultation of the records fails to disclose any failures of material in service that have been traceable to the failure of accepted material under normal shock conditions. However, impact tests should be the principal criterion of any shock-resisting material, and it is hoped that their use in both commercial and ordnance work will rapidly come into favor.

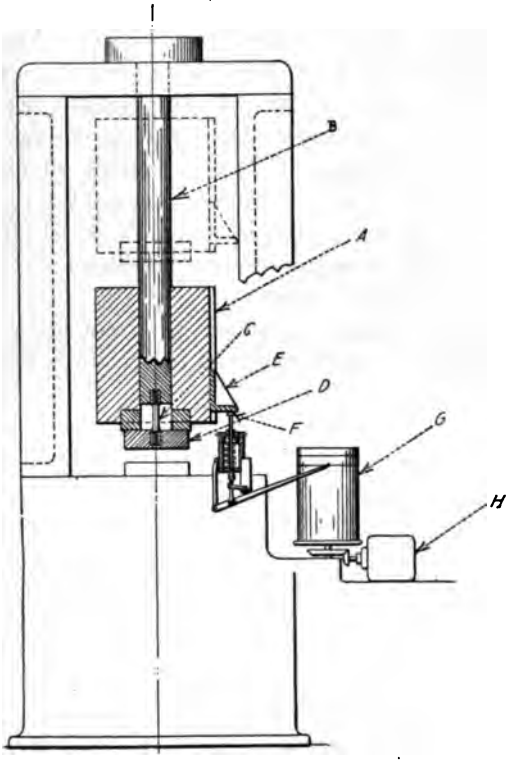

FIG. 22.—VERTICAL DROP MACHINE.

It is considered imperative, however, for the general use of such a test, that conditions should be standardized in the beginning, so that no confusion or change of equipment will result upon the incorporation of a standard test in the specifications. It is considered that there should be an early decision as to standard American practice in the following particulars, in order that the impact test may become a part of the routine procedure of buying materials subject to sudden stresses:

1. A selection of either the pendulum or the vertical drop type of impact testing machines.

2. A selection of either notched or unnotched bars.

3. If notched bars are selected, the immediate selection of a standard notch, so that the practice of playing with various types of notches, such as has been done in other countries, be nipped in the bud, and some consistent data will quickly be assembled.

4. If possible, the adoption of an impact test whereby the resistance of the material can be reduced to an equivalent of pounds per square inch, which will have a more definite meaning to the ultimate consumer of this country and will obviate the necessity of a tabulation of new units in the customary reports of physical test.

This last point is not so far removed from possibility as it may seem at first glance. What appears to be a very workable arrangement has been devised at the U. S. Naval Gun Factory, for use with a vertical drop machine, by Lieut. G. Q. Lewis, of the Navy, who was assigned this task in connection with the contemplated activities of the enlarged physical laboratory of the Gun Factory. The essential features of the apparatus consist of a weight *A*, Fig. 22, which is guided by a central shaft *B* and can fall from any predetermined height, striking a hardened plate *D* and rupturing in tension the standard test bar *C*, the ends of which are screwed into the lower end of the shaft *B* and the plate *D*, respectively. A bracket *E* suitably secured to the weight *A* engages a plunger *F* of a device that amplifies the vertical movement of the weight just before and during the period of rupture. A pencil record of the amplified movement is made upon a paper cylinder mounted on a drum *G*, which is rotated at a known

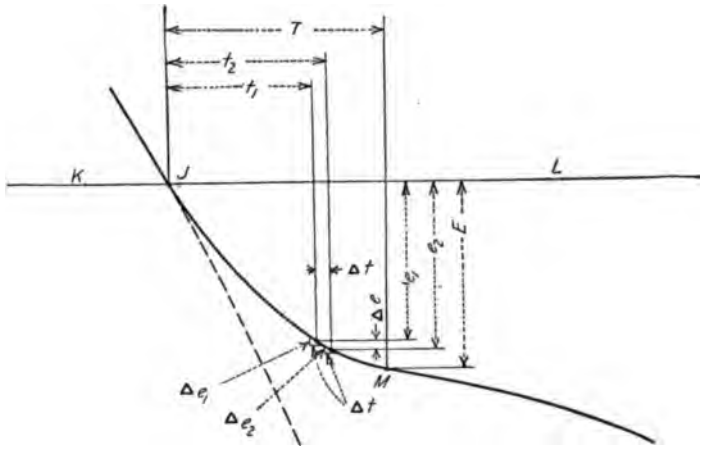

FIG. 23.—RELATION OF ELONGATION OF TEST BAR WITH REFERENCE TO TIME.

constant speed by the motor *H*. The object of the amplification is to obtain a space-time curve of such proportions that a graphic or protractor analysis may be conveniently and accurately made and the resisting value of the material tested derived in terms of pounds per square inch.

The curves obtained show the relation of the elongation of the test bar with reference to time. For a typical curve (see Fig. 23), the elongation would be measured vertically and the time horizontally to known scales, depending on the amplification and surface speed of the drum. The weight, being released from a suitable predetermined height, accelerates uniformly and attains maximum velocity at a certain point *J* where the pencil line crosses a horizontal base line *KL*, previously scribed on the drum with the weight resting on the plate *D*, Fig. 1, which is screwed to the lower end of the test bar. The tangential motion of the paper drum

being to the left, the rupture curve is drawn downwardly and to the right. Above the base line is recorded the last portion of the parabolic free fall of the weight and below is recorded the movement of same as retarded by the resistance of the test bar. The curve reverses after crossing the base line and the slope of the same decreases as the bar elongates, owing to the reducing velocity of the weight. After rupture occurs at M, the curve reverses and the weight again accelerates, due to gravity.

Analysis of the curve between the initial and the final points of test-bar resistance gives the average forces of retardation corresponding to increments of elongation, Δe, measured normally to the base line. To determine the average retarding force during any time increment Δt, the slopes or tangents of the curve are first obtained at the beginning and the end of the time increment Δt, to determine the corresponding instantaneous velocities of the weight at these points. Thus the tangent to the curve or its equal, the instantaneous velocity of the weight at the beginning of time increment Δt, is $\frac{\Delta e_1}{\Delta t}$ and at the end is $\frac{\Delta e_2}{\Delta t}$. The acceleration during the period Δt is $\frac{\Delta e_2 - \Delta e_1}{\Delta t^2}$.

The average force of retardation F, in pounds, during any increment of time Δt, therefore, equals the product of the mass and the acceleration, or $K\frac{\Delta e_2 - \Delta e_1}{\Delta t^2}$, where K is a constant depending on the mass of the weight and the units employed in measuring Δe and Δt. F will be algebraically negative because the acceleration is negative. Thus the stress exerted by the test bar to resist the retardation force of the weight W will be $+F$, since it is equal but opposite in direction to that imposed by the weight. The total stress in the bar will then be algebraically equal to $W + F$, where W is the number of pounds in the weight, also resisted by the test bar because of gravitation.

It is not to be assumed that calculations of this description will be necessary for each specimen pulled, as a protractor has been designed which will integrate the space-time curve throughout the entire elongation period of the specimen and thus permit a ready selection of critical values, read directly in pounds per square inch.

While Lieutenant Lewis, who has now returned to his former activities in the supplying and testing of railroad materials, had the benefit of previous research in connection with railroad shock absorbers, it is considered that this method of analysis is considerably different from any performed before, in that it goes further and gets an expression of the strength of material, in pounds per square inch, which is a function of the amount of energy expended in rupturing the specimen.

The Naval Gun Factory will appreciate receiving criticisms of the proposed method, and, when the new physical laboratory is operating, will be glad to share with the members of this Institute any information obtained on this question.

JOHN H. NELSON,* Worcester, Mass.—At our plant we have done considerable work on notch-bar testing and are farther at sea now than when we started. So far, we have been unable to get the notch-bar test to be consistent with any other known test. Lately, I have been investigating the effect of the position of the notch-bar specimen in a bar. To do this I took a 3½ in. square bar and quartered it, then heat treated the four quarters. Each of these pieces was again quartered and a specimen machined and notched. In this way we would have notch-bar tests from the exterior to the interior of the bar. These were made of such lengths that it was possible to make four impact tests on each specimen.

Extreme care was taken in marking the position of these test specimens in the original bar. The notches were then cut with relation to position of the specimen in the original bar. The second notch was cut 90° from the first, and so on around the specimen. In this way, we have found that, with a 3½ in. bar of 0.4 or 0.5 per cent. carbon steel when properly heat treated, we could get impact values ranging from 15 to 100 ft.-lb. Just why this is the case we are unable to state; but it does appear to be a fact that along the outside of the original bar the notch-bar test seems to be low and increases toward the center, but it is not always consistent. For instance, we may have, on a single specimen, one notch-bar test giving us 15 ft.-lb. and another 100 ft.-lb.

We have not found such inconsistencies in alloy steels. Alloy steel is consistently good or consistently bad. If it is good, you cannot readily destroy the impact value, no matter what you do; if it is bad—that is, if it is a poor notch-bar steel—and you have sufficient latitude in your draw, you may correct it. In other words, if the hardness you are working to will allow you to draw to about 1050° F.(561° C.) or over and then quench this steel from the draw heat, you can raise the impact value of the steel from about 4 or 10 ft.-lb. up to 40 or 50 ft.-lb. Just what occurs I do not know. There is evidently a critical point in steels that have poor impact value, in the neighborhood of 1000° to 1050° F. (536° to 561° C.).

I have taken a number of coupons of chrome-nickel steel with low impact value and heat-treated as follows: Quenched all coupons in water from the same temperature, then varied the draw temperature on these coupons by 50° and water quenched each coupon from the draw heat. No increase in impact value was found until a drawing temperature of between 1000° and 1050° F. was used, when the impact value of the steel was greatly increased over that obtained when the same draw temperature was used and air-cooled.

I have also taken the same steel, water-quenching this steel from the draw heat to produce good impact values, and then redrawing these

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coupons beginning with about 500° F. and varying the draw temperatures by 50°, and air-cooling them in every case to determine at what point the impact values would be reduced by air-cooling. This method again indicates that the critical point is somewhere in the neighborhood of 1000° to 1050° F. The unnotched impact tensile test bears absolutely no relation to the notched-bar test. By using an impact tensile specimen which is not notched, equally good results may be obtained on low notch-bar steel or a high notch-bar steel. There is no choice between the two.

We have adopted as an impact tensile specimen a diameter of about 0.235 in. and a 1 in. gage length. This will give results on elongations which are comparable with those obtained on the standard 0.505 in. diameter and 2 in. gage length. By using this specimen, better results on elongations and contractions may be obtained on the impact tensile test than upon the static tensile test.

We have been unable as yet to determine the exact meaning of the notch-bar test when used in comparison with the commonly accepted tests on steels. We are continuing to work on this most interesting subject and trust that sooner or later we may be able to determine the true meaning of the notch-bar test.

F. C. LANGENBERG (author's reply to discussion*).—An enormous tonnage of forgings from which a transverse tensile test cannot be taken, due to the dimensions of the forgings themselves, is produced annually. An example is an automobile axle. Because a forging is small is no indication that it will not be subjected to stresses in a transverse direction. Although it is true that the forgings reported on by the author were of such size that both transverse and Charpy tests could be taken, the forgings had to be selected in order to establish the relation pointed out in the paper.

It is not necessarily cheaper to machine a tensile bar than a square-notched impact bar, provided proper equipment is available for the impact bar.

We do not feel entirely at sea regarding the significance of the notched-bar test. It was the purpose of the paper to point out one of the principal relations between this test and certain fundamental properties of the material being tested.

* Received Jan. 20, 1920.

Does Forging Increase Specific Density of Steel?

BY H. E. DOERR,* ST. LOUIS, MO.

(New York Meeting, February, 1919)

THE writer has been unable to find much information relative to tests made to determine the effect of forging on the specific density of steel. The opinion, however, among men engaged in the business, seems to be that forging greatly compresses or consolidates the steel. While such is doubtless the case with spongy or porous steel, the following experiments indicate that there can be little or no change in the density with steel initially free from cavities.

Ten ingots of basic open-hearth steel cast as shown in Fig. 1 were used in the experiments. All the specimens were 0.53 carbon with all other elements the same, the analysis being determined from drillings taken midway between the center and outside of the upper end, as shown in Fig. 2. The specimens were selected from over one thousand ingots so that the tests could be comparative for the work done in forging only, but no two of the ingots were cast in the same heat. Each ingot remained in the mold approximately 30 min., when it was removed and air-cooled. About 12 hr. later, when the temperature had been reduced to approximately that of the surrounding atmosphere, the head was discarded and the ingot heated in an oil furnace to approximately 2250° F. (1235° C). Approximately 1 min. was required to forge under the steam hammer to 1¼ in. (31.75 mm.), rough diameter, as shown by the dotted lines in Fig. 2. All ingots were allowed to air-cool after forging. Cylinders ½ in. (17.7 mm.) in diameter and ½ in. high were then turned from the forged and unforged parts of the ingots and numbered. The ten pairs of specimens were submitted to Louis E. Endsley, Professor of Railway Mechanical Engineering, University of Pittsburgh; A. N. Talbot, Professor of Applied Mechanics, University of Illinois; L. Z. Slater, chief chemist, Scullin Steel Co., St. Louis, Mo. No information was furnished as to which specimens were forged or unforged and each laboratory used the same specimens, submitting the average of three independent determinations, a tabulation of which is shown in the accompanying table in which the laboratories are designated by the letters A, B, and C. While there is considerable difference in the individual determinations for a number of the specimens, the averages

*Chief Mechanical Engineer, Scullin Steel Co.

for both the forged and unforged specimens are very close. The determinations of each laboratory indicate that the forged is the denser of each pair with the single exception of "B's" determinations for No. 10. Assuming that all cavities, however small, were completely closed, the average results indicate that only 0.64 per cent. of the sectional area of the original ingot consisted of cavities.

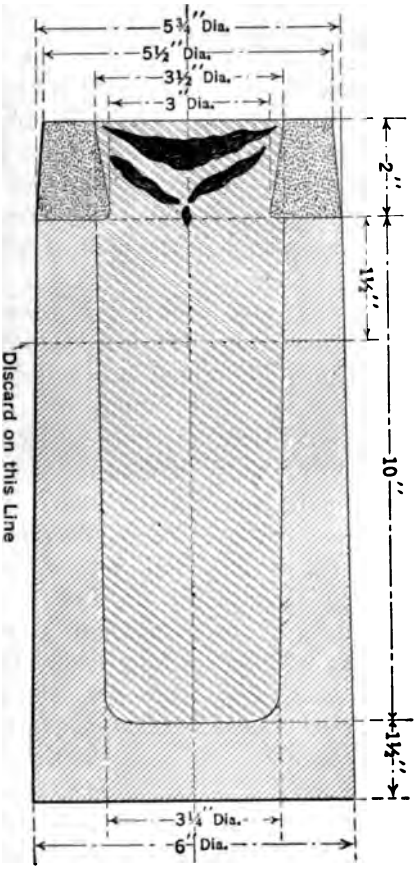

FIG. 1.—BASIC OPEN-HEARTH STEEL INGOT.

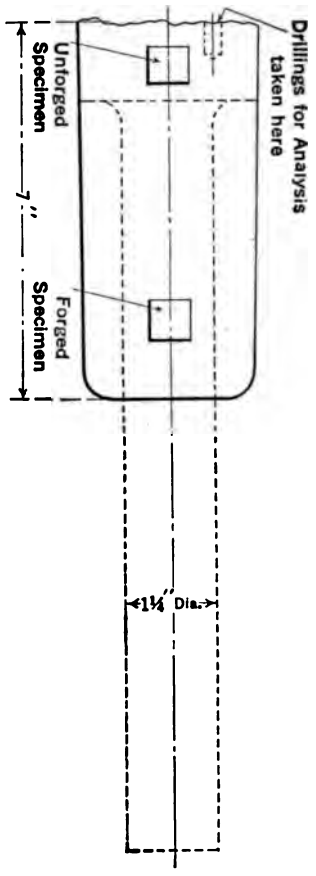

FIG. 2.—FORGED INGOT SHOWING POSITION OF SAMPLES.

The ratio of volume to superficial area is so great with an ingot of the size used that it is doubtful whether the ingot can properly solidify without the formation of microscopic cavities. Undoubtedly, with large ingots where the ratio of volume to superficial area is smaller than with the ingots used, the percentage of increase in specific density of the forged specimens would be reduced to a negligible value, if not wholly eliminated.

Effect of Forging on Specific Density of Steel

Specimen	Density Determined by Laboratory				Specimen	Density Determined by Laboratory			
	A	B	C	Average		A	B	C	Average
No. 1 forged	7.807	7.86	7.826	7.831	No. 7 forged	7.808	7.84	7.840	7.829
No. 1 unforged	7.732	7.81	7.815	7.786	No. 7 unforged	7.694	7.77	7.791	7.752
No. 2 forged	7.825	7.85	7.824	7.833	No. 8 forged	7.828	7.85	7.863	7.847
No. 2 unforged	7.756	7.79	7.792	7.779	No. 8 unforged	7.759	7.79	7.809	7.786
No. 3 forged	7.810	7.85	7.818	7.826	No. 9 forged	7.818	7.84	7.854	7.837
No. 3 unforged	7.779	7.79	7.781	7.783	No. 9 unforged	7.755	7.78	7.798	7.778
No. 4 forged	7.837	7.84	7.875	7.851	No. 10 forged	7.820	7.85	7.873	7.848
No. 4 unforged	7.774	7.80	7.790	7.788	No. 10 unforged	7.807	7.86	7.835	7.834
No. 5 forged	7.828	7.86	7.829	7.839	Average of forged	7.821	7.849	7.845	7.838
No. 5 unforged	7.802	7.83	7.800	7.811	Average of unforged	7.762	7.798	7.803	7.788
No. 6 forged	7.830	7.85	7.846	7.842	Increased density of				
No. 6 unforged	7.761	7.76	7.814	7.778	forged, per cent	0.76	0.65	0.54	0.64

DISCUSSION

J. S. UNGER,* Pittsburgh, Pa. (written discussion†).—There is a common impression in the minds of many that forging or rolling a piece of steel not alone increases the strength but makes the steel more dense, or causes it to become heavier. I believe it is the author's aim in the paper to show that cast steel is almost as dense as forged steel. The average results obtained by him show very little difference in density, being but 0.64 per cent. between the unforged and forged specimens. Some steel castings are used in an unannealed condition but the usual practice is to anneal them. As the author heated the ingots to a temperature of 1235° C. for forging, and after cooling cut specimens from the forged and unforged portion, the results are really a comparison between steel annealed at 1235° C. and steel forged from the same temperature. The usual annealing temperature for castings is about 850° C., or about 385° C. less than the temperature used in the experiment.

To secure some information as to the density of steels as cast, annealed, and forged, and to show the effect of composition, the following experiment was made. Small ingots 3 in. (7 cm.) in diameter by 10 in. (25 cm.) long were cast in a dry-sand mold from four different heats of steel and were allowed to go cold. A piece 3 in. long was cut from the bottom of each ingot, and from the upper end of this piece duplicate specimens for density were cut. The remainder of piece was then annealed for 2 hr. at the temperatures shown in the accompanying table, allowed to cool slowly, and duplicate specimens cut. The upper part of the ingot was heated to 1150° C. and forged to 0.75 in. (19 mm.) diame-

* Manager, Central Research Bureau, Carnegie Steel Co.

† Received Feb. 18, 1919.

ter in 1½ min., allowed to cool in the air, and duplicate specimens cut from the lower end, or what was originally adjacent to the first specimens. This furnished duplicate specimens as cast, as annealed, and as forged.

The results are shown in the table. The first two steels are heaviest in the annealed state and lightest after forging; the last two are heaviest as cast and lightest as forged. Also the density is shown to increase as the carbon increases; this is not in accord with some other investigators, but may be accounted for, in part, by the high silicon in the last ingot, which contained 55 per cent. carbon. There is very little difference in the percentage in any set of specimens from the same heat, but the forged specimens are lighter, or have a lower specific gravity, than either the cast or annealed specimens.

Specific-gravity Determinations of Steel Ingots

Compositions					Annealing Temperature, Degrees Centigrade	Specific Gravity as Cast	Specific Gravity as Annealed	Specific Gravity as Forged
C	Mn	P	S	Si				
0.11	0.53	0.012	0.041	0.026	900	7.817	7.832	7.809
0.35	0.53	0.014	0.044	0.047	850	7.827	7.828	7.824
0.49	0.52	0.016	0.050	0.033	850	7.833	7.832	7.828
0.55	0.41	0.015	0.046	0.250	850	7.862	7.859	7.850

G. AERTSEN.—I would like to ask whether those results are from the averages of specimens, or just the result of one specimen?

J. S. UNGER.—They represent one specimen only, there are duplicate specimens from each piece. The work I had done in the past on the same subject indicated that the higher the carbon, the lower the specific gravity, but I was very much surprised to find in these results, the higher the carbon, the higher the specific gravity.

H. M. HOWE,* Bedford Hills, N. Y.—Here, as in so many cases, we have two factors at work which produce opposite results. We only get the mean of those results in such an investigation as this. An increase in carbon content means the substitution of the lighter cementite for the denser ferrite. This consideration should cause the density to decrease as the carbon content increases. On the other hand, as the carbon content increases the tendency to form gas cavities decreases, so that this effect of carbon would cause the density to increase with the carbon content. In short, an increase in carbon content has two distinct results of which one, the increase in cementite, tends to lower the density, whereas the other tends to increase the density. Hence in one

* Chairman, Engineering Division, National Research Council.

set of observations you may find that the density increases and in another that it decreases as the carbon content increases.

As regards forging you have very much the same thing. The act of forging tends to close up any gas cavities, of course; on the other hand, as many know from bitter experience, the act of forging tends to cause internal axial cavities or pipes, as in the Mannesmann process. This tendency to form internal axial cavities by forging has to be guarded against with very great care. It was a great and, in one case I believe, a fatal obstacle to making objects of varying cross-section such as axles and projectiles by rolling them between dies, as in the Simonds process. And so one set of investigations may lead in one direction and another lead in the opposite direction, unless great care is taken to distinguish between these two opposite results, opposite purely and simply mechanical results of one and the same process, namely, forging.

Static, Dynamic, and Notch Toughness

BY SAMUEL L. HOYT,* E. M., PH. D., MINNEAPOLIS, MINN.

(New York Meeting, February, 1919)

SOME of the more important properties of finished materials are strength, ductility, toughness, resistance to alternating and repeated stresses, etc. Of these, the property that appears to have received the least consideration, at least in this country, is toughness, which is due, possibly, to our regarding toughness as a qualitative property or, at any rate, as one that parallels ductility. This arises out of our greater experience with the simple carbon steels in the rolled or forged (non-heat-treated) condition which, qualitatively, are tough if ductile and are not tough if brittle. In this way we have erroneously become accustomed to judging the toughness of a material from its behavior in the tensile or similar test.

The point of view presented in the present paper is that toughness, like hardness or tensile strength, should be regarded as an independent property and of sufficient importance to require, in so far as that may be possible, quantitative determinations. If such be true, it becomes at once necessary to devise experimental means for measuring or valuating toughness, and the notched-bar impact test (the Charpy test) is advanced as the most logical test that has so far been developed for this purpose. It is further advanced, with particular emphasis, that we have two kinds of toughness to deal with and, accordingly, they will be dealt with independently.

CLASSIFICATION OF MATERIALS ACCORDING TO TOUGHNESS

Toughness has been defined somewhat as follows: Tough materials are those that offer considerable resistance to permanent deformation but which, once such resistance has been overcome, may be deformed plastically, but only by the expenditure of considerable energy. In other words, tough materials may be deformed plastically but they absorb a considerable amount of work in the process. This kind of toughness may be called "static" toughness when the rate of loading is reasonably slow or "dynamic" toughness when the rate of loading is comparatively rapid, as in impact testing, but in all cases the stress distribution is essentially uniform. However, static toughness does not

* Associate Professor of Metallography, University of Minnesota.

imply resistance to shock, or dynamic toughness. In fact, dynamic toughness may be equal to, greater than, or less than the static toughness, thus dividing materials into three classes. This is well shown by numerous cases on record.

Tests on the resistance to impact of cast-iron bars, as made by Russell,[1] indicate that cast iron is nearly one-half again as resistant under impact as it is under static loading, using for the comparison the amounts of work absorbed in producing fracture. Considere,[2] working on the dynamic resistance of soft-iron wire, showed that to produce a given deformation greater loads were required if suddenly applied than if slowly applied, but that the total deformation (ductility) remained constant. According to this, soft iron absorbs more work when broken by dynamic loading than by static loading.[3] One of the most notable cases of this kind is that described by De Fréminville[4] in discussing the application of impact testing to the selection of metals for use in machine construction. De Fréminville considered two types of parts according to their behavior under impact, both of which must be made of materials that are highly resistant and able to absorb heavy impacts. The first type must do so and yet not deform appreciably, while the second type is allowed considerable deformation provided the part so stressed is able to retain its original shape when the load is released. During the course of this work, one steel was found that was considerably more resistant under impact than under static loading, and so, for his purpose, was particularly valuable. The ordinary (static) tests failed to bring out the superior quality of this steel. The type of materials that has practically the same behavior under static and dynamic loading is fairly large, as has been shown by the work of Breuil,[5] Hatt,[6] Frémont,[7] Charpy,[8] and Martens.[9]

Tests on the third type of materials, which possess remarkably low resistance to impact, were carried out by Considere[10] who brought out

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[1] S. Bent Russell: Experiments with a New Machine for Testing Materials by Impact. *Trans.* Amer. Soc. Civ. Eng. (1898) **39**, 237.

[2] Contrib. à l'Étude de la Fragilité dans les Fers et les Aciers, 3.

[3] This, of course, does not mean that the material can stand repeated loads greater than the known tensile strength even though they be applied suddenly.

[4] Contrib. à l'Étude de la Fragilité dans les Fers et les Aciers, 475.

[5] Pierre Breuil: Abstract of paper on effects of stresses. *Jnl.* Iron and Steel Inst. (1904) **65**, 413.

[6] W. Kendrick Hatt: Tensile Impact Tests of Metals. *Proc.* Amer. Soc. Test. Mat. (1904) **4**, 282.

[7] Contrib. à l'Étude de la Fragilité dans les Fers et les Aciers, 150.

[8] Ibid., 213.

[9] Adolf Martens: "Handbook of Testing Materials," translation by G. C. Henning, 233. John Wiley & Sons, New York, 1899.

[10] *Loc. cit.*

some very interesting points in connection with the dynamic properties of metals. The resistance to impact of soft-iron wire under shear and under tension was measured with varying velocities of impact by gradually increasing the height of drop of the weight producing the impact. It was found that the resistance to impact increased directly with the velocity of the impact up to and even beyond that velocity which was sufficient to cause the rupture at one blow. By still further increasing the height of drop, a velocity was reached at which the material broke suddenly with very low impact resistance and with negligible deformation. At velocities above this critical velocity, the resistance to impact is obviously much inferior to the static strength. By increasing the weight to a certain (critical) amount, the resistance to impact decreased sharply as soon as the velocity was sufficient to cause the rupture with one blow.[11] In such a case, heavy impacts are much more dangerous than light impacts even though the amount of energy expended is the same in both cases. Temperatures from $-10°$ C. to 150° C. were found to have only a minor (primary) influence, but a very important secondary temperature influence was found, inasmuch as low temperatures caused a marked lowering of the critical velocity. These tests served also to bring out the inferiority, under impact, of hard-drawn iron wire as compared to annealed wire. Under static loading, the tensile strength of the unannealed wire was greater than that of the annealed wire, but under impact both the deformation and the tensile strength were less for the unannealed wire, which shows that in certain cases a high static tensile strength is not a sufficient guarantee of strength.

Notch Toughness

It is well known that a stress applied to a bar that has a sudden change in cross-section along its length produces a decidedly non-uniform strain distribution at the change in cross-section. If the change in cross-section is in the form of a nick or a groove, the strains at the base of the nick multiply and are much greater than the average strain over the cross-section. Such a nick, or sudden change in cross-section, is here referred to as a "notch," and the non-uniform strain distribution, as the "notch effect." The ability of a material to withstand stresses when in the notched condition is referred to as its "notch toughness."

The notch effect is well illustrated every time a blacksmith nicks a bar to break it off at any particular point. Even a blow by the hand produces strains at the base of the notch well in excess of the resistance of the material and hence produces the fracture. A similar blow on an

[11] The work of Considere suggests the advisability of considering all impact testing (weight and velocity of impact) from the point of view of the critical velocity and critical weight here described.

unnotched bar would merely bend the bar over. Thus it is that a bar, even though made of normally tough material, if notched, may behave as if brittle. Koenigsberger[12] has shown by means of glass models that notches localize the strains and that the neutral axis of a stressed bar runs close to the peak of the notch instead of remaining in the middle section. Heyn[13] has shown that a lead bar cut with a notch has an entirely different strain distribution under bending from that of a similar bar without a notch. The volumes of the strained parts were as 1 : 3.76 and the maximum fiber elongations, measured by the distortion of 5-mm. squares, were 120 per cent. and 70 per cent. for the notched and unnotched bars respectively. But to get a correct idea of the maximum deformation, the extension of the width of a line at the apex of the notch was determined. The original width of the line was 0.25 mm. but after deformation it was found to be 5 mm., which gave a deformation of 1700 per cent. It has also been shown. by Leon,[14] that a transverse notch cut in a tensile test bar produces an uneven distribution of strains over the cross-section at the base of the notch, the greatest strains coming at the apex of the notch. These cases illustrate the notch effect as it occurs in ordinary metals, which is to localize the strains in the neighborhood of the apex of the notch, at times to such an extent that they may be far in excess of the resistance of the material under tension. In excessively brittle materials, the multiplication of the maximum fiber stress is even more excessive. An example of a notch in such materials is the small file scratch made on a glass rod to assist in breaking off any particular length.

It is likewise known that the severity of the notch effect increases as the angle of the notch decreases. On this account the notch effect increases in most materials when fracture starts, because the angle of the fracture is generally less than the angle of the original notch. Certain pliable materials, or those that are "self healing," of which lead is an example, behave in the opposite manner, since the notch becomes more rounded with distortion. In the same way a material of high notch toughness is much less dangerously affected by a notch than one with low notch toughness, a point that will receive consideration further on.

Notch Effect in Engineering Practice

The frequency of the notch effect in engineering practice has been brought rather forcibly to the writer's attention in various investigations of failed parts. It was noted that the notch effect may at times be

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[12] *Proc.* 4th Congress Intern. Assoc. Test. Mat., Brussels, 1906, Paper C 4 d.

[13] Martens-Heyn: "Handbuch der Materialienkunde." II A, 374. Berlin, 1912. Julius Springer.

[14] *Oester. Wochenschrift f. d. oeffentlichen Ba[illegible]*

intentionally introduced through the design of the part (sometimes faulty) and at other times may be unintentionally introduced through faulty or careless workmanship; of these, the latter is the more reprehensible and the more difficult to guard against. Another equally important point is the necessity of considering the microstructure and the desirability of heat treatment as a means of overcoming or counteracting the effect of the notch.

A certain railway company suffered an extraordinary number of splice bar failures, which from the records it was impossible to trace to roadbed conditions. Tests on parts of the failed bars showed that the material was well up to the quality required by the specifications, so that no basis for criticism could be made on that score. It contained 0.34 per cent. carbon

FIG. 1. × 65.

FIG. 2.—SAME AS FIG. 1. × 425.

ORIGINAL STRUCTURE OF SPLICE BAR.

and 0.018 per cent. phosphorus; it had an elongation of 35 per cent. on 2 in. (50.8 mm.); a tensile strength of 65,200 lb. (29,574 kg.); and a yield point of 33,000 lb. (14,968 kg.). On making a microscopic examination, the steel showed considerable free ferrite, in characteristic Widmannstättian structure, as may be seen from Figs. 1 and 2. Furthermore, the pearlite was of the familiar lamellar type that is characteristic of slowly cooled steels. The idea developed from this examination was that the two rails and the splice bars form a notch and that the structure of the bar was such that it could not always adequately resist the notch effect produced every time a car passed over the rail joint (repeated stresses). The solution of this problem obviously lay in correcting or improving the structure of the splice bar, to which end a series of heat-treatment tests was conducted. By subjecting the bar to a temperature of 900° C. for ½ hr. and quenching in oil, a structure was secured that was principally sorbite with only a small amount of free ferrite. The tensile strength

was increased to 92,000 lb. (41,730 kg.) and the yield point to 45,000 lb. (20,411 kg.) but the elongation was decreased 22.5 per cent. A characteristic structure is reproduced in Figs. 3 and 4, which represent a small piece of a failed splice bar heat-treated in the laboratory. The original coarse Widmannstättian structure is replaced by a network structure,

FIG. 3. × 65.

FIG. 4.—SAME AS FIG. 3. × 565.

STRUCTURE OF SPLICE BAR AFTER HEAT TREATMENT.

the major part of which is sorbite, a constituent composed of the original pearlite and most of the original ferrite. Compared to the Widmannstättian structure, sorbite may be said to be highly resistant to the notch effect. By the adoption of heat-treated splice bars the difficulty was eliminated.

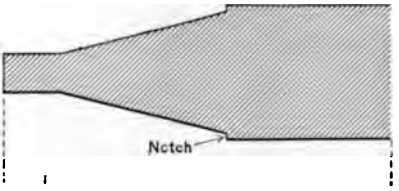

FIG. 5.

Some truck forgings, solid rear axles and steering arms, failed while in heavy service. These parts were made by a well-known automobile axle manufacturing company and suspicion rested at first upon the truck drivers, particularly as the material was known to pass all specifications. It contained 0.485 per cent. carbon, had an elongation of 27.1 per cent. in $3\frac{1}{2}$ in., a reduction in area of 45.6 per cent., a tensile strength of 77,500 lb. (35,153 kg.), and a yield point of 39,900 lb. (17,690 kg.). An examination of the axle (not heat-treated) showed that the taper had

been cut by a roughing tool in such a way as to leave the notch indicated in Fig. 5. Knowing the danger of the presence of such a notch if the steel were in a poor physical condition, a microscopical examination was made; this showed the condition represented in Figs. 6 and 7. Here again the

FIG. 6. × 65.

FIG. 7.—SAME AS FIG. 6. × 565.

ORIGINAL STRUCTURE OF REAR AXLE.

presence of a relatively large amount of excess ferrite will be noted. The result of a simple oil quench of a part of the failed axle is shown by Fig. 8. While the free ferrite has been largely eliminated and sorbite has been substituted for pearlite, an even better heat treatment would be

FIG. 8.—STRUCTURE OF AXLE WHEN OIL QUENCHED. × 130.

to quench the axle in water, to entirely prevent the segregation of ferrite, and to reheat to produce sorbite and the requisite mechanical properties.

The danger of a keyseat in an axle is well illustrated by Figs. 9 and 10, which show the extensions of two cracks leading from the angles of the

keyseat. If good engineering practice calls for such a keyseat, the material should be suitably heat-treated to give it a high notch toughness; or, better yet, the material itself, as well as the heat treatment, should be selected with reference to its notch toughness.

An examination of the truck steering arms (which were heat-treated) also revealed the presence of a notch due to defective design or construction of the drop-forging die. Microscopic examination of the failed steering arms showed the heat treatment to have been imperfect inasmuch as granular pearlite, instead of sorbite, was formed which, while

FIG. 9.—LEFT SIDE OF KEYSEAT. × 16.

FIG. 10.—RIGHT SIDE OF KEYSEAT. × 16.

ductile and strong, is not at all a satisfactory structure in the presence of a notch. A similar examination of satisfactory steering arms showed sorbite.

TESTS FOR TOUGHNESS

It has been customary to test materials by various static and dynamic tests and to judge from the strength and ductility whether the quality of the material is sufficiently high to warrant its use regardless of the design of the part, its relationship to other parts, or possible defects due to faulty workmanship. One of the objects of the present paper is to show that, in case the material is to be used in the notched condition, the usual tests are unable to differentiate clearly between materials that will probably stand up and those that are likely to fail. This means that it is necessary to supplement the ordinary tests with a test on notched bars.

In order properly to define what is meant by toughness, it was necessary to distinguish between conditions of uniform and of non-uniform strain distribution. Likewise in testing the toughness of materials the same distinction must be made, a point that, due to its more or less

general disregard, is somewhat strongly emphasized here. A qualitative test for toughness quite commonly used is the nick and fracture test for fiber. This test is useful to a certain extent but it should have no more place in scientific testing of materials than an ordinary scratch test for hardness. The reduction of area in the tensile tests is very frequently looked at with an idea of sizing up the toughness of the material. It is more correctly a measure of ductility and becomes a measure of static toughness only in a limited sense, inasmuch as it is the ability to deform and not the resistance to such deformation that is determined. The area of the stress-strain diagram may be taken as a measure of the static toughness. Such measurements show, for example, that the toughness of annealed carbon steels increases with the carbon content up to about 0.3 per cent. carbon. This figure, combined possibly with the resistance to impact, should be useful in case the strain distribution of the finished part is essentially uniform. The value obtained by dividing the tensile strength by the proportional limit and multiplying by the elongation (Martens) may be very useful in certain cases; for example, in bringing out the toughness of pure copper. The value of this determination is also limited, as it does not include the resistance with which a material opposes permanent deformation. All of these values are of undoubted importance when properly interpreted, but none would bring out the weakness of the materials in the foregoing examples. In fact, due to their apparent indication of strength and resistance, they would be directly misleading.

The recognition of the peculiar weakness of certain materials when in the notched condition has led to the development of the impact test on notched bars as a supplement to the customary static and dynamic tests and for the express purpose of testing materials for toughness. In its present form, the test is the result of systematic experiments extending over two decades or more by Barba, Frémont, Charpy, Ast, and others, which culminated in the reports of Charpy to the International Association for Testing Materials in 1909 and 1912[15] and of Ehrensberger to the German Society for Testing Materials in 1907.[16] The accumulation of evidence over this period made possible the establishment of a standard test so that now it can be safely stated that the notched-bar test, or in particular the Charpy test, is capable of supplying information relative to toughness that the tensile test gives in but an imperfect, and often in a directly misleading, manner. The test shows the great danger of angular notches or sudden changes in cross-section, particularly when material of low noteh toughness is used.

[15] G. Charpy: Report on Impact Tests of Metals. *Proc.* Int. Assn. Test. Mat. (May, 1908–Feb., 1910) **1**, No. 5, III_1; Report of Impact Tests and the Work of Committee No. 26 (May, 1910–May, 1913) **2**, Pt. 2, IV_1.

[16] Ehrensberger: *Stahl u. Eisen* (1907) **27**, 1797.

In the report of Ehrensberger, results of tests on three classes of materials were given: forged carbon steels, forged special steels, and cast carbon steels. These three classes of materials can be compared by means of the figures shown in Table 1, which are taken from the report.

TABLE 1.—*Properties of Three Kinds of Steels*

Test No.	Tensile Strength, Pounds per Square Inch	Yield Point, Pounds per Square Inch	Elongation,* Per Cent.	Reduction in Area, Per Cent.	Notched Toughness, Meter-kilograms per Square Centimeter	Remarks
	Forged Carbon Steels					
1	61,500	32,700	26.5	64	4.6	Forged too hot.
2	64,100	36,100	26.0	70	20.4	Correctly forged.
9	71,600	42,000	24.5	70	22.6	
10	71,800	40,000	26.4	60	4.7	Failed railway axle.
25	142,200	93,000	12.1	36	8.5	High carbon.
	Forged Nickel and Chrome-nickel Steels					
51	72,800	58,200	23.3	70	42.1	
69	130,000	108,600	15.1	62	22.1	
71	142,200	118,200	13.3	56	19.3	
75	270,200	232,500	6.5	31	8.3	
	Cast Steel					
161	68,000	37,700	22.9	51	3.7	

* The elongation is for a gage length equal to 10 times the diameter.

A few comparisons may serve to bring out the value of the Charpy test for toughness. Tests 1 and 2 show that the notched-bar test brings out the lack of toughness (or at least notch toughness) of test bar 1, which was forged too hot, although no evidence of this was given by the tensile test. Tests 1 and 25 show very plainly that the toughness cannot be entirely judged from the reduction of area or elongation; test bar 1 is the more ductile of the two but test bar 25 possesses greater notch toughness. Test 10 shows how far the tensile properties can come from indicating lack of toughness as the probable cause of failure, although the true character of the material is brought out by the notched-bar test. A comparison of the carbon steels with the special steels reveals a superiority for the latter that is not as clearly brought out by the tensile tests. The tensile tests indicate that the special steels have greater tensile strength for the same ductility but the superior toughness

of the latter is better brought out by the notched-bar test. The tensile properties of the cast-steel specimens show that cast steel may have excellent elongation and reduction of area, and be thus apparently ductile, but be quite lacking in resistance when tested in the notched condition. Thus cast steel behaves the same as a piece of over heated steel.

In Charpy's report of 1909, an interesting case was cited to show that the notched-bar test can give information regarding steel that is in no wise suggested by the tensile tests, either static or dynamic. The results obtained with two steels A and B, really the same steel in two

TABLE 2.—*Results Obtained by Notched-bar Tests*

Steel	Elastic Limit, Pounds per Square Inch	Tensile Strength, Pounds per Square Inch	Elongation, Per Cent.	Reduction of Area, Per Cent.	Static Resistance, Kilograms	Tensile Impact Resistance, Kilograms	Charpy Test, Kilograms
A.....	42,800	59,900	32.0	67.2	179.5	205	44.0
B.....	42,700	62,700	32.0	65.6	185.0	195	2.7

different conditions of heat treatment, are given in Table 2. Not even the resistance to fracture produced by a weight falling from a height of about 100 ft. (30 m.) as given in the next to the last column, indicates the excessive brittleness of steel B. The microstructure at once showed that B is in a much poorer condition than A.

These figures could be multiplied almost indefinitely and have been well known for over a decade. There can now no longer be any doubt that there is some property, which is of great technical importance, that is not measured and in many cases not even indicated by the usual tensile test. In spite of this undoubted fact, the Charpy test, which is now extensively used in Europe, is given but little attention in this country. This would seem to be due to lack of familiarity on the part of users of steel with the fact that in a very large number of cases materials are used in a notched condition, whether intentionally or otherwise, and that when so used the tensile test offers no reliable index of their probable behavior. If these points were considered, the purchaser would certainly insist that his materials show a high degree of notch toughness in all cases where they are to be used in the notched condition.

Various objections have been raised to the adoption of the notched-bar test, one of the principal ones being the so-called lack of uniformity of the results. This question has been gone into rather thoroughly by H. M. Howe,[17] and more recently by Charpy and Cornu-Thénard.[18]

[17] *Trans.* (1913) **47**, 501.

[18] G. Charpy and A. Cornu-Thénard: *Rev. de Mét.* (1917) **14**, 84; *Jnl.* Iron and Steel Inst. (1917, No. 2) 61.

In a number of cases the results of the notched-bar impact test are not as concordant as might be desired. It is the opinion of many experimenters, however, that variations in the impact resistance of supposedly similar test bars are due to actual variations in the material, the wide scale of the Charpy test throwing them into greater prominence than do the usual tests. The truth of this contention is demonstrated by the recent work of Charpy and Cornu-Thénard, who, by using exceptionally uniform and homogeneous bars, secured check results as close as 1 to 2 per cent. and scarcely ever varying as much as 4 per cent.[19] The results obtained by the commission of the German society were sufficiently concordant to lead to the adoption of the test and the minor variations were not permitted to mask the fact that the notched-bar test is capable of yielding valuable information that the tensile test does not.

The writer is informed that at least one steel plant, which has done considerable work on the notched-bar test including the Charpy test, has found the concordance of results to be very satisfactory and capable of yielding possible differences of impact resistance of the order of magnitude of 1:30. Thus, it would seem that lack of concordance can now no longer be advanced as an objection to the adoption of the Charpy test.

There has also been considerable objection to the presence of the notch in the test bar, the idea being that the nick localizes the breaking point and so does not test the weakest cross-section.[20] This is undoubtedly a point that must be considered when using the Charpy test. For example, if there is excessive segregation it is quite possible that the base of the notch will come at the weaker part and so lead to low results. At the same time such a variation in results is a valuable indication of segregation, so that by breaking a number of test bars not only can the average value be obtained but non-homogeneity in the material itself can be determined. The object of the Charpy test is not so much to test any particular test bar,[21] as it is to test the relative toughness of a certain material[22] or the efficacy of a certain heat

[19] Even a 4 per cent. variation could scarcely be objected to on the score of lack of concordance in the light of the present practice in pulling test bars. A 4 per cent. variation would mean a variation from 49,000 to 51,000 lb. per sq. in., a variation that is much less than variations due to improper rate of loading, improperly holding the test bar in the machine, and in particular to the general neglect of the proportional limit in favor of the yield point as a measure of the elastic limit.

[20] H. M. Howe: The Resilience Test. *Met. & Chem. Eng.* (1917) 17, 298; Walter Rosenhain: "Introduction to the Study of Physical Metallurgy," 237. New York, 1915. D. Van Nostrand Co.

[21] The use of the Charpy test as an acceptance test would hardly merit discussion until the value and importance of the test proper is generally conceded.

[22] This statement is subject to the reservation advanced by Professor Howe that, due to the greater plasticity of low-carbon steels, we cannot directly compare steels of essentially different carbon contents.

treatment. Moreover, it is quite customary, at least in certain quarters, to precede the Charpy test by macroscopic tests for gross segregations in order that variations due to such segregations may be eliminated.

Recently there has been an attempt to eliminate the notch from impact testing, as in the impact shear test described by Doctor McAdam. In this test, an unnotched bar is sheared by an impact while the impact figure is read the same as in the Charpy test. It has been argued[23] that by eliminating all except one stress the impact shear test is superior to the Charpy test. There can be no doubt that materials to be subjected to shearing stresses by impact (in an unnotched condition) might well be tested by this method; but no test on unnotched bars can be substituted for a notched-bar test. The correctness of this assertion is supported by the results of the impact shear test according to which such materials as properly heat-treated nickel and chrome-nickel steels fail to give as much resistance to impact as carbon steel or even ingot iron.

Summary

Undoubtedly the occurrence of the notch effect in machines and engineering structures is much more common than is generally recognized. This is generally due to the design of the structure but may be caused by faulty workmanship. Even a hasty examination of such machines as locomotives, automobiles, stationary gas engines, steam engines, etc. reveals an amazingly large number of notches and in many such cases the material composing the parts should have a high degree of notch toughness to insure against failure.

The logical test for such materials and the only one capable of yielding reliable results is the notched-bar test. This test should supplement the usual tensile or hardness tests and its results used as an index of the resistance of the material to the notch effect.

A factor of safety is generally used in the design of parts of machines requiring the material to have a certain strength combined with a certain amount of ductility. These properties are written into the specifications and the material is inspected on such a basis. Neither the factor of safety nor the usual properties offer a guarantee against failure in cases similar to those discussed here. It is for this reason that the cause of many failures remains a mystery when test bars taken from the broken parts are found to pass all specifications. The Charpy test would undoubtedly show a low impact value, due either to faulty heat treatment, in case the parts were heat treated, or to lack of proper heat treatment.

[23] Personal communication.

DISCUSSION

J. A. MATHEWS,* Syracuse, N. Y. (written discussion†).—This paper and the papers by Messrs. Jeffries,[1] Clayton, Rawdon, and Moore,[2] constitute a valuable symposium on what might be called the "mechanism of failures." Professor Hoyt has not emphasized too strongly the importance of the notched-bar test, or the notch as a feature of a structural design. There has been a good deal of misunderstanding in reference to the value of the notched-bar test, and various vibratory tests such as the Wohler, White-Souther, Stanton, and Upton-Lewis have been devised. This form of testing has received more attention abroad than it has here, and has been the subject of a great many very careful investigations. I have heard many expressions of surprise at the apparent lack of agreement between the results of vibratory tests and the results of shock tests. It is not necessary that the values should agree, because different properties are being tested.

In reference to the presence of notches or sharp corners in the construction of machines or parts, I would say that these more frequently lead to fatigue failures than to typical notch-bar failures due to sudden impact. I have seen a number of airplane crankshafts that had apparently failed due to a keyway that was left with fairly sharp angles at the bottom. The material satisfactorily responded to every test, including the notched-bar test, nevertheless the shafts failed, showing a fatigue break, usually spiral and sometimes making at least one complete turn around the shaft.

The lack of concordance between supposedly duplicate notched-bar tests in my experience has been most often due to lack of care in the machining of the test piece itself. Professor Hoyt refers to a report of Charpy in 1909, in which two steels of almost identical physical properties show very different results in the notched-bar test. He states that the microstructure at once showed that one steel was in poorer condition than the other. More recent experience has shown other cases of this kind where not even the microstructure serves to reveal any startling differences in structure.

The notched-bar test serves admirably to detect faulty heat treatment, as well as to detect defects in the material itself. The typical alloy steels for the most part have a fairly wide range within which heat treatments can be performed without showing any marked differences in the ordinary tensile qualities. When the temperature becomes too high there is apt to be a falling off in elongation and reduction, but not very marked. The differences in the shock-resisting qualities, however, are much greater than would generally be supposed from examination of the tensile-test data.

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* President, Halcomb Steel Co.
† Received Feb. 18, 1919.

[1] Effect of Temperature, Deformation and Grain Size on Mechahical Properties of Metals. *Trans.* (1919) **60**, 474.

[2] This Volume, 211, 246, 397.

The true function and differences between vibratory and notched-bar tests are very clearly pointed out in a letter from Dr. T. E. Stanton, of the National Physical Laboratory, a copy of which was furnished me by Sir Robert Hadfield. Dr. Stanton says: "In the first place it may be mentioned that the determination of the fatigue range of stress will involve generally about five or six separate tests on similar specimens prepared from the same bar. For the preliminary test a range of stress is chosen which will cause the failure of the specimen after about, say, 100,000 reversals. For the succeeding tests, this range is gradually reduced until the maximum range is found for which no failure will take place for however long the test is carried on. I think there can be no doubt that this maximum range of stress for an unlimited number of repetitions is the most valuable strength characteristic of the material which a designer of structures or machines can possess.

"In the second place, it must be fully recognized that the test does not attempt to discriminate between brittle and tough materials, which is, of course the function of impact testing. Considerable criticism has been passed on the fatigue tests because material which is known to be brittle often gives a high value of fatigue range of stresses. This criticism, however, is based on an imperfect knowledge of the limitations of the test."

Please note the expression Dr. Stanton uses, namely, "the fatigue range of stress." In fatigue tests, so far as they have been considered for the most part in this country, we have considered a single value or determination by any of the numerous methods of vibratory tests as an indication of quality or properties in itself, just as we would speak of a Brinell hardness number. The use which Dr. Stanton proposes for this test is apparently quite different. He states the true function of the fatigue test to be "the determination of the maximum working stress to which material may safely be subjected in practice," and he presupposes that the material has a satisfactory microstructure and that the toughness, as revealed by the notched-bar impact test, is sufficient for the purpose for which the material is to be used.

Professor Hoyt's paper deals primarily with external notches. In my judgment, an even more important consideration is internal notches in the form of flakes, slag inclusions, bursts, pipes, and all kinds of interior defects. The external ones can be seen and avoided; the internal ones are more difficult to detect. In this connection, I would like to call attention to a most interesting paper by Dr. Andrew McCance, entitled "Non-metallic Inclusions and Their Constitution and Occurrence in Steel," which was presented at a meeting of the Iron and Steel Institute of Great Britain in May, 1918. I will quote the opening paragraph of this paper, which is very striking:

"The important part which non-metallic inclusions play in causing

failures and producing defects in all manner of steel products is not yet fully realized. Much defective steel is bad, solely because of the number of non-metallic particles which it contains, and fully 90 per cent. of the failures due to faulty material which have come under the author's notice have been traceable to this cause alone."

There is food for much thought in that paragraph, and subsequent discussion by the author, showing how these inclusions may be and are a potential source of failure, is worth serious study. The most minute of these inclusions may lead to what have been designated as hair lines. These are so minute that they only appear on the ground and finished surface, and often require a lens for their detection. The number of them present in different steels varies tremendously. In a series of tests made by grinding and inspecting large surfaces, the relative frequency of hair lines was about as follows: In basic open-hearth steel an average of 46 hair lines was noted; in acid open-hearth steel, the average was 23; and in electric steel, the average was 6. In a great many electric heats that I have seen tested in this way, no hair lines whatever were detected. Not only the number, but the relative size of the hair lines was very different in the electric from those seen in the open-hearth steels; they were very much shorter and less conspicuous, as well as less frequent.

It is my opinion that if designers were fully acquainted with the materials they were using they could design much lighter parts, because they would be able to design with reference to a fairly definite factor of safety, whereas at present they mostly employ a generous factor of ignorance. It has taken the very serious jolt of the war to bring about the intensive study and coöperation in regard to these problems which we see today, and this has brought about an appreciation of high quality in materials that did not formerly exist.

PAUL D. MERICA,* Washington, D. C. (written discussion†).—Professor Hoyt has well emphasized the danger of failure in ferrous materials of construction that attends the simultaneous presence of sharp offsets in section, or notches, and a thoroughly crystalline structure of the net work type. This combination is found to be dangerous also in non-ferrous materials. Thus in an investigation made several years ago of a number of failures of brass and bronze[24] a number of naval brass rivets were examined, some of which had stripped off at the head during several months, others of which, however, had remained intact. It is obvious that the offset at the shoulder or head of the rivet forms a sharp angle at which tensile stresses on the rivet are severely localized, and those rivets that had broken, broke at this point. The photograph shows the

* U. S. Bureau of Standards. † Received Feb. 7, 1919.

[24] P. D. Merica and R. W. Woodward: Failure of Brass. U. S. Bureau of Standards *Tech Paper* 82 (1917).

structure typical of those that broke; the metal consisted of grains of the beta surrounded by complete envelopes of the alpha constituent. The fracture was intercrystalline and along these envelopes. The majority of those rivets that had not broken, but which had been removed for examination, showed a quite satisfactory and fine structure free from gross network.

I have, elsewhere,[25] called attention to the significance of the notch in determining the phenomenon of corrosion cracking in brass and bronze. It appears that corrosion is much more severe at the bottom of sharp notches or offsets in such materials than at other parts, when the article is under tensile stress, and may actually determine failure at such points if the stresses are severe. H. S. Rawdon[26] has corroborated this fact in the course of some very interesting studies on the corrosion of Muntz metal.

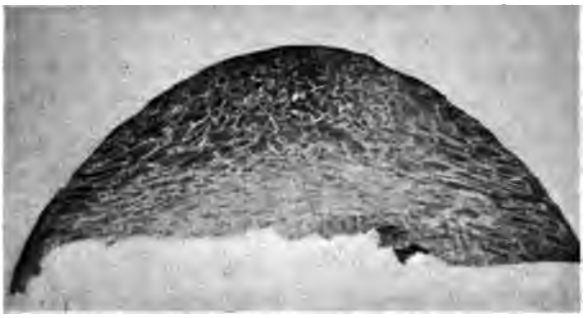

FIG. 11.—ETCHED SECTION OF NAVAL BRASS RIVET FRACTURED IN SERVICE.

GEORGES CHARPY,* Paris, France (written discussion†).—We have read with interest Professor Hoyt's paper on the toughness of steels considered from the viewpoints of static or dynamic tests, and of notch tests. Professor Hoyt has ably presented the question, illustrating with most convincing photographs, and metallurgists will thank him for his work.

Influence of the Speed of Impact.—The writer believes that one should guard against exaggerating the importance of the influence of the speed of test. It seems to have been proved, by numerous experiments, that when the test is kept within practical limits the speed does not exert a preponderating influence. The results of the static test differ much from the results of the dynamic test, but the proved differences

[25] P. D. Merica: Failure of Brass. U. S. Bureau of Standards, *Tech. Paper* 83 (1916).

[26] Typical Cases of the Deterioration of Muntz Metal by Selective Corrosion. U. S. Bureau of Standards *Tech. Paper* 103 (1917).

* Member, *Institut de France.*

† Received Feb. 13, 1919. Translated from the French.

do not exceed a few units per cent. in the most favorable conditions; that is, they are of the order of magnitude of the differences observed from one experiment to the other in the same series.

About 15 years ago, the writer made some tests on resistance to shock in which the ram (drawing with it the resisting bar) fell about 50 m., with a speed of impact of about 30 m. per sec. The energy absorbed by the rupture in these conditions did not differ notably from the energy absorbed in the course of the test on the static machine (length of test = 1 min.) calculated by interpolating a curve. For the notch tests, the question seems to present itself in the same manner. Attention was attracted 20 years, or more, ago to the factor of speed by the experiments of Considère, but the results of these seem to have been falsified by an error of interpretation, and it would seem wise not to take them into account.

Influence of Treatment of Materials on Results of Impact Test.—The author has very ably brought forward the importance of effect of the treatment undergone by the steel on the manner in which it behaves in an impact test; it is an observation that the writer has often made. A metal, even if it is not of exceptional quality, being given the factors of safety ordinarily used, if it is well treated, can furnish pieces that will give excellent results. On the contrary, if metal is not treated at all, or is treated in a wrong way, even though it may be sound and unalloyed, it will furnish very unsatisfactory pieces. The impact test on notches seems to the writer valuable for discovering this absence of treatment, or defects in the treatment because, of all the ordinary tests, it is the only one that shows these points with exactness.

That is the reason that the writer believes, with the author, that the localization by the notch in a specified section is an essential condition of the impact test. It is not necessary to seek the local faults that may be shown by other methods, but only to point out precisely the physical state as a whole of the metal constituting the piece to be tested. It is therefore necessary to make the test on a deliberately chosen region, and to shield oneself as much as possible from the local accidents that falsify the results as a whole.

Definition of Notch.—The author describes judicially the influence of the shape of the notch on impact tests; the writer believes that it would be a considerable advantage in comparing experiments if the notches were rigidly standardized. All observers, at the present time, seem agreed in recognizing that the bar of 10 by 10-mm. section is the best adapted to the current practice of the workshop; the lengths employed vary scarcely more than from 53.3 to 60 mm. But we find, in scientific literature, an amazing diversity in the shape of the notches; some use the sharp-pointed superficial notch; others, the deep round notch; and all intermediate types are represented. How, under these

conditions, can we compare the results obtained in one laboratory with those of another; the results of one experimenter with those of another?

After a long series of tests and of practical observations, the writer believes that the best notch is the deep one (mid-thickness), and round. It seems to present the following advantages: (1) It can be determined with the greatest exactness, since it can be obtained with the drill. (2) Like all deep notches, it reduces in the whole work of rupture, the proportion of the work of bending in reference to the work of detaching the surfaces, this last being the most interesting in the phenomenon of fragility. (3) Its making does not affect the physical state of the metal around it (hardening undergone by the steel in the case of notches made by the wedge).

In the writer's opinion, a great advantage will be gained when the different experimenters adopt one style of notched bar. As noted in the preceding, the writer favors a bar 10 by 10 by 60 mm., with a round notch 3 mm. in diameter and 5 mm. deep. This notch was recently defined by a sub-committee specially appointed by the French Commission for the Unification of Terms of Reference. This sub-committee was composed of M. Mesnager, Chief Engineer of Bridges and Roadways, Director of the Laboratory of Bridges and Roadways in Paris; M. Cellerier, Director of the Testing Laboratory of the Conservatory of Arts and Trades in Paris; and M. Charpy, General Secretary of the Commission for the Unification of Terms of Reference.

J. E. JOHNSON, JR., New York.—It seems to me that it ought to be somebody's job to tie these notch tests with the reversed-stress tests, because the thing that causes failures in structures, in a great majority of cases, is that the material is stressed repeatedly up somewhere near the elastic limit, and the repetition of that stress seems to bring about what I suppose you may call a growth of crystals and that eventually develops into a break. It seems to me that it would be of the greatest value if there could be a coördination of the results of a shock test of some kind with the reversed-stress test, so that we could be sure that a material that had shock-test resistance would have reversed-stress resistance.

H. F. MOORE,* Urbana, Ill.—The question raised by the speaker concerning the correlation of results from impact tests of notched bars with the results of repeated-stress tests is a very complicated one. Impact on a specimen sends energy into the specimen, and if the specimen absorbs this energy without fracturing there must be developed both stress and elongation. A notch in the specimen allows but little elongation to take place before rupture, and develops the ductility of the

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material only at the particular section through the notch. In a repeated-stress test, failure starts at some irregularity of surface outline on the specimen or at some internal defect. This irregularity of outline—such as a sharp tool mark—or internal defect acts like a small notch, but the reason for failure at such a point is quite different from the reason for failure of a notched specimen under impact. Mathematical analysis, direct tests, microscopic examination, and results in practice all indicate that at the root of a small notch or deep scratch the localized fiber stress may be several times as great as the stress computed by the formulas in ordinary use. We ordinarily do not pay any attention to these little localized stresses because they affect only a small area and do not cause any measurable increase of general deflection of a machine or structural member. However, under repeated stress, such a localized stress may start a crack, which is in itself a deeper notch, and thus the damage spread. I do not know of any correlation of repeated-stress tests with impact tests, I do not believe that we can assume that such a correlation is possible. Such a correlation may be discovered in the future, and if clearly proved would be very important.

J. E. JOHNSON, JR.—My idea was to correlate the notch tests with the reversed-stress test. However, by leaving the test piece plain, you eliminate the advantage that Mr. Moore has spoken of.

JOHN H. NELSON,* Worcester, Mass.—In reply to the question "Is there any final information in regard to the relation between the temperature or the rate of cooling of a piece of heat-treated steel from the drawing temperature, and the so-called notch toughness?" I might say that we have conducted a large series of tests in the study of impacts on chrome-nickel and carbon steels and have gone into the subject of quenching from drawing temperature quite thoroughly but have found that one cannot lay down any definite rule or law with respect to the effect of quenching from the drawing heat. We conducted experiments on the effect of cooling by quenching in water, cooling in air, and burying the coupon in lime. Steels that have practically no impact value when air cooled may show very marked results when they are quenched in water, depending on the temperature at which the drawing is done.

There appears to be a critical temperature to which the steel must be heated before any effect is obtained. For the particular steel on which we have been working, this point is located somewhere between 1000° and 1050° F. (538° and 566° C.). If the steel is quenched at a temperature below 1000° F., no useful result whatever is obtained; if it is quenched from a temperature above 1000° F., a very marked result is produced. A steel that has a good impact figure apparently is not affected, regardless of whether it is quenched from the drawing temperature or is air cooled

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or slowly cooled; that is, cooled in lime. Just what the difference is in the steels we are unable to say. Nearly every steel that has caused us any difficulty on impact tests has been reported to us as acid; while the steels with which we have had no difficulty have been reported basic. We have had some acid steels, however, that gave good impact test.

With regard to the statement that the effect of heat treatment on the steel was shown markedly by the impact figure, we have tests to prove that this is not true; also tests that prove that it is true. That has been our general experience with the impact test; we can present figures to prove anything we desire. We have been trying to determine what the impact tells us. So far it hasn't told us anything; that is, the story is not consistent. It will tell us one thing one day and the next day it will tell us something entirely different. We made a series of tests, using a 1450° F. (788° C.) quench and varying the draw, and then with the same steel we used a varying quenching temperature up to 2000° F. (1094° C.) with a constant drawing temperature, and in both cases have obtained excellent results with both the tensile and the impact test. With the same steel and with a 50° lower drawing temperature, entirely different results were obtained.

When our investigation is completed we expect to publish the results obtained.

Oxygen in Cast Iron and its Application

BY WILFORD L. STORK,* M. E., A. M., WYANDOTTE, MICH.

(Chicago Meeting, September, 1919)

CERTAIN influences of oxygen on iron have been known for many years and it has always been considered one of the worst enemies of the iron and steel founders. Nobody had a good word for it, hence little was done to make use of this oxygen. In all the fields of foundry practice, the desideratum was "melt without oxidation of the metal."

In the consideration of the influence of oxygen on cast iron, we must not confuse this condition with oxidized iron caused by poor melting conditions, due to a high bed, heavy charges, low coke ratio, excessive melting rate, high blast, or poor charging. It is admitted that oxygen absorbed by iron melted under these conditions has a deleterious effect on the properties of the iron. The writer has had experience both in the melting under such poor conditions and in remedying these bad conditions and considers the product far inferior to iron melted under normal conditions. As stated by Dr. Richard Moldenke,[1] such iron has a higher melting point and high combined carbon, is sluggish, and contains slags and oxides in solution in the liquid, causing slag inclusions and gas occlusions, thus producing defective castings. This oxidized condition cannot be counteracted or removed by any of the known deoxidizers; it can be remedied only by proper melting conditions. Such iron is really burned, whereas the iron considered in this paper was melted under normal conditions with moderate blast, correct bed height, light charges of iron, and low coke consumption and is not in this category at all.

In 1914, the late J. E. Johnson, Jr., startled the metallurgists of the world when he published his results on the advantages of oxygen on cast iron.[2] His work is so convincing and so logical, in spite of the common and well-established notions on the subject, that nobody has been able to refute a single argument. Such an eminent authority on iron and steel metallurgy as Prof. H. M. Howe has accepted the work in all its details in his latest book.[3] The work of J. E. Johnson, Jr., has revolutionized the metallurgy of cast iron and it seems apparent that all future work must include an oxygen determination just as we include silicon, sul-

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[1] "Principles of Iron Founding." New York, 1917. McGraw-Hill.

[2] *Trans.* (1914) **50**, 344.

[3] "Metallography of Steel and Cast Iron." New York, 1916. McGraw-Hill.

fur, phosphorus, manganese, and carbon in our every-day analyses. A summary of the part of this work that applies directly to the contents of this paper is as follows:

1. Iron containing oxygen solidifies at a higher temperature than iron of the same composition containing no oxygen.

2. Oxygen increases the strength of cast iron.

3. Iron containing oxygen, if cast in sand, contains somewhat more combined carbon than a similar iron not containing oxygen.

4. Iron containing oxygen, if cast against a chilling surface, will show a white chill at a higher silicon than a similar iron not containing oxygen, and will show a greater chill for the same silicon as the latter element decreases.

5. When manganese is added to a metal high in oxygen, some of the latter is removed and the chilling power and strength are thereby reduced.

6. Steel, when added to cupola mixtures, is oxidized and it is this oxygen content that gives such mixtures its superior physical properties.

The purpose of this paper is to give further evidence in support of the above theories and to show their application to cupola mixtures.

Iron poured hot will always have more combined carbon in the test bar or casting than the same iron poured colder by holding it in the ladle; this is true for both white cast iron and gray iron. It is possible to get a white, mottled, or gray fracture of white cast iron, simply by pouring at different temperatures, in accordance with the well-known law that the more slowly iron is cooled the more graphitic and less combined carbon will the iron have. But when iron is held in a boiling ladle (first catch in a relined ladle), the heat of the molten iron drives off the water of combination from the clay, which when converted into steam passes through the iron causing boiling. The steam is decomposed into hydrogen and oxygen; the oxygen attaches itself to the iron molecules and the hydrogen passes off to the atmosphere. The result is the iron becomes less fluid, not merely due to its lowering temperature, and the combined carbon is increased as shown in the following analyses:

	Series A		Series B	
	Poured Hot, Per Cent.	Poured After Boiling, Per Cent.	Poured Hot. Per Cent.	Poured After Boiling, Per Cent.
Combined carbon	0.40	0.64	0.28	0.44
Graphitic carbon	2.85	2.96	3.08	3.04
Manganese	0.81	0.70	0.82	0.82
Phosphorus	0.309	0.301	0.340	0.316
Sulfur	0.113	0.0604	0.096	0.083
Silicon	1.98	1.90	2.04	2.03

In series A, the manganese and the sulfur are reduced in the proportion of 2:1, the theoretical proportion to form manganese sulfide; also the silicon is oxidized. The iron in the state of boiling gave the manganese and sulfur the opportunity and time to get together to form manganese sulfide, which rises to the surface as slag. The oxygen content of the iron increased its combined carbon and raised its melting point, causing it to become sluggish.

In melting much iron of different analyses, day by day, the adjustment for the manganese loss varied in the different mixtures, but was quite constant in each particular grade. This had to be considered in order to keep the mixtures up to specifications. The mixtures carrying steel lost a greater proportion of manganese than those without steel For instance, in a mixture of 50-per cent. pig iron and 50-per cent. remelt, the manganese content was:

	Charged	Spout	Loss	Loss, Per Cent.
Manganese	80	69	11	13.7

whereas in a cylinder mixture made up of 15-per cent. steel, 40–50 per cent. remelt and 45–35 per cent. pig iron, the manganese content was:

	Charged	Spout	Loss	Loss, Per Cent.
Manganese	90	71	19	21.1

With special mixtures, as will be shown later, the loss of manganese is considerably higher. These are the results of averages over a period of months of operation. The steel in the cupola oxidizes, adding oxygen to the mixture, which in turn is removed by the manganese. The percentage of manganese lost increases with the increased steel content of the mixture, for, the higher the steel content the greater is the oxidation and the more oxygen to be removed. In the cylinder mixture carrying 15 per cent. steel, it was necessary to charge the iron at 0.90 per cent. or above in manganese, in order to give the iron proper fluidity to run such complicated castings as automobile cylinders. The oxygen content due to the steel raised its melting point. Of two irons at the same temperature, the one with the lower freezing point will have more superheat and more life than the one with the higher freezing point. The oxygen added to the iron through the steel raises its melting point, causing it to become sluggish; and in order to give the iron the required life, it must be deoxidized with manganese. This fact was so readily observed that fluidity was controlled entirely by manganese.

Iron with 20 per cent. steel chills more easily, has a greater shrinkage, is harder, and will throw a deeper chill than iron with 10 per cent. steel, unless extra manganese is added to remove this additional oxygen.

With the advent of the United States in the great world war, this country followed the experiences of France, Italy, and Germany in speci-

fying cast iron for its 155-mm. gas shells. In order to meet the Government's severe specification for this material, the writer applied the aforesaid theories and facts to his mixtures and had very good success. The iron was deliberately impregnated with oxygen by heavy percentages of steel and deoxidation was prevented by keeping the manganese below 0.45 and always above 0.33 per cent. The Government demanded a tensile strength of over 32,000 lb. per sq. in. (2249 kg. per sq. cm.) and an impact test. The impact test required that a 1¼-in. (32-mm.) square test bar supported on knife edges, 6 in. (152 mm.) center to center, should not break before the 25-lb. (11-kg.) hammer, carrying a knife edge, reached 18 in. (46 cm.) starting at 12 in. (30 cm.) and increasing by ½ in. (12.7 mm.) increments. Mixtures were made of 35–40 per cent. steel, 35 per cent. remelt, and 30–25 per cent. pig iron. A typical analysis and set of physical properties of two heats are: 150 lb. Star pig iron, 450 lb. Toledo pig iron, 700 lb. remelt, 700 lb. steel, 2 lb. ferromanganese.

Silicon, Per Cent.	Sulfur, Per Cent.	Phosphorus, Per Cent.	Manganese, Per Cent.	Combined Carbon, Per Cent.	Graphitic Carbon, Per Cent.	Total Carbon, Per Cent.	Chill, Inch	Transverse, Lb. per Sq. In.	Impact, Inch	Brinell Number
1.33	0.121	0.194	0.45	0.78	2.42	3.20	0.10	3300	22.5	420
1.35	0.121	0.208	0.45	0.81	2.39	3.20	Trace	3400	24.0	420

The tensile strength ran from 32,000 to 37,000 lb. per sq. in.; bars which broke below 32,000 lb. proved to be defective castings.

The following results bear out the various contentions in reference to the oxidation of steel in cupola mixtures and the effect of oxygen on

Silicon, Per Cent.	Manganese, Per Cent.	Combined Carbon 1½ × 1½-in. Bar, Per Cent.	Chill 1 × 5 × 3 Inch	Silicon, Per Cent.	Manganese, Per Cent.	Combined Carbon 1½ × 1½-in. Bar, Per Cent.	Chill 1 × 5 × 3 Inch
1.18	0.27	1.51	0.90	1.35	0.44	0.78	0.35
1.18	0.29	1.47	0.80	1.35	0.48	0.75	⅛corner[a]
1.16	0.30	1.27	1.05	1.36	0.50	0.70	1/16 cor.[a]
1.17	0.35	1.17	0.75	1.32	0.53	0.58	Trace[a]
1.16	0.38	0.94	0.30	1.43	0.34	1.49	0.75
1.15	0.40	0.94	0.65	1.45	0.35	1.18	0.60
1.17	0.50	0.76	0.20	1.43	0.39	1.01	0.45
1.18	0.51	0.69	⅛corner[a]	1.45	0.43	0.69	0.25
1.36	0.31	1.53	?	1.43	0.47	0.64	0.15
1.35	0.35	1.32	0.75	1.49	0.33	1.45	?
1.34	0.37	1.20	0.50	1.49	0.33	1.23	0.40
1.34	0.38	1.05	0.50	1.48	0.42	0.75	0.20
1.35	0.41	0.94	0.22	1.49	0.53	0.68	0.05

[a] 2 in. × 2 in. × 6 in. chill test.

cast iron. For like silicon contents, the combined carbon and chill are decreased as the manganese is increased, due to the manganese acting as a deoxidizer. These results are not for the same ladles of iron but are from different days and are the general trend of conditions and results.

The manganese loss was very high in these mixtures, as shown by the following, with which the losses under less oxidizing conditions are repeated for comparison.

Per Cent. Steel	Manganese			
	Charged	Spout	Loss	Per Cent. Loss
0	80	69	11	13.7
15	90	71	19	12.1
35	64	38	26	40.0

There was an abnormally large amount of slag in these runs and it became necessary to reduce the limestone from 80 to 20 lb. (8 to 9 kg.) per ton of iron, then it was found possible to discontinue the use of limestone with these heavy steel charges. Without any limestone the slag was sufficiently fluid and no trouble was experienced from a viscous slag due to the fact that there was sufficient FeO to flux with the SiO_2 and a good running slag was produced. The slag analyses proved the high oxidation of the steel. Below is given the analysis of a slag made from the shell mixture using no limestone as compared with those of slags made from 0- and 10-per cent. steel, using 80 lb. of limestone per ton of iron.

	0-Per Cent. Steel	10-Per Cent. Steel	35-Per Cent. Steel
SiO_2	42.0	44.3	52.78
Al_2O_3	20.2	22.7	21.60
CaO	30.4	26.4	2.88
MgO	3.0	2.6	4.45
FeO	1.3	2.5	13.60
MnO	1.6	1.8	3.64

The increase in FeO and MnO in the increased percentages of steel proves the conditions in the melting of the iron and corroborates its effects on the iron. Also with 0- to 10-per cent. steel, the slag hole was opened only after 2 to 2½ hr. of operation whereas with these special steel charges it was advisable and almost necessary to slag the cupola 30 to 40 min. after the first tap.

The contents of this paper proves the validity of the late J. E. Johnson, Jr.'s theory of oxygen in cast iron and shows how it has been applied to cupola mixtures to obtain iron of definite physical properties and also how it explains many conditions met with in foundry practice. Even

the chill can be controlled by varying the manganese in irons containing oxygen and the influence of oxygen on the fluidity can be controlled by the manganese. The oxygen content can be controlled directly by steel additions and indirectly by varying the manganese. While this method is not as good as the patented process of oxygenating cast iron, nevertheless we can make use of the method indirectly and can apply the theory involved to our daily foundry problems. It also shows the absurdity at times of considering the influence of silicon on the chill, strength, or fluidity of cast iron when the oxygen content is neglected, as we can produce almost any desired chill for any silicon content simply by varying the oxygen.

DISCUSSION

R. MOLDENKE, Watchung, N. J. (written discussion*).—It is somewhat difficult to discuss the paper of Mr. Stork, when the description of the cupola melting occurrences indicates that his practice is open to question. The evident burning of much of the steel charged, coupled with the remarkably low silicon content of the mixture as shown by the resulting product, throws the entire melting practice into that of making cupola malleable. If anything is wanted here, it is a high bed to melt upon, and yet Mr. Stork in the very beginning of his paper speaks of the evils of oxidation due to a "high" bed, etc. The experienced foundryman would judge at the outset that Mr. Stork has kept his bed in the cupola too low, and naturally got a lot of oxidation from the blast.

The oxidation of steel when used in large quantities in cupola mixtures is a well-known phenomenon, and is readily understood. One has but to put a file into the forge fire and watch it burn when heated high enough. The melting point of steel being higher than cast iron, heavy pieces may pass far into the bed before being completely melted away, the last portions being directly exposed to free oxygen from the blast. Another situation not to be forgotten is the oxidizing effect of carbon dioxide at high temperatures. A serious oxidation of the steel is quite conceivable if in the zone of maximum carbon dioxide and minimum carbon monoxide—free oxygen being entirely absent. Hence the well understood practice of never melting a mixture with much steel unless the bed is kept exceedingly high; or again the above-mentioned malleable practice.

Mr. Stork mentions the impossibility of deoxidizing a heat after the steel contained has been oxidized. This is true to a great extent with the ordinary deoxidizers, such as silicon and manganese, but can be brought about by vanadium, titanium, magnesium, etc. Proper melting, however, makes this unnecessary.

Mr. Stork differentiates between the oxygen taken up by melting steel

* Received June 30, 1919.

and that by ordinary mixtures under improper melting conditions. It is difficult to understand this contention. To the experienced foundry metallurgist, there seem to be but two forms of iron oxide to be reckoned with in cupola melting. One is the rust or surface oxidation particularly prevalent in such material as stove-plate scrap, and the other the oxide actually within the metal, as in "burnt" grate bars, soda kettles, furnace plates, etc. Rust is simply taken up in the slag, but the other form enters the molten metal and more or less of it remains dissolved therein, depending upon the opportunity given to react with manganese, silicon and carbon. This depends in turn upon the temperature and the mobility of the molecules. Unless it is hot enough to allow of a reaction and the mixing is intimate enough, oxygen and carbon, in fact manganese and sulfur, will in the common iron vehicle flow out side by side over the cupola spout into the ladle, get into the mold and actually sometimes get together while in the mold, with disastrous results to the machining qualities and soundness of the castings.

No matter what may be brought up in the way of photomicrographs of oxygenated iron, etc., there does not appear to be the slightest effectual difference between this and a burnt soda-kettle iron, when both have been melted again. Indeed, the instance cited by Mr. Stork, of the boiling action of green ladles, does not differ from Johnson's oxygenation. If continued too long it is bad, but in slight measure the higher combined carbon produced yields a stronger metal. The foundryman knows this and on occasion uses the homely method of spitting a potato on an iron rod and poking it down into the ladle, with beautiful boiling effects.

To take up the six points mentioned by Mr. Stork as summarizing the oxygen effect on cast iron, the following may be said: The fact that cast iron containing oxygen solidifies at a higher temperature than cast iron free from this element was probably first pointed out by me many years ago and continually dwelt upon, so that now it is generally recognized as true. Every foundryman knew that what he called "burnt" iron set faster and had little "life." He did not know that dissolved oxide of iron caused this, scientifically speaking, raising of the "freezing point." It is with considerable satisfaction that I now see my years of contention for the recognition of oxygen in cast iron accepted by eminent metallurgists who scouted the idea until the work of the late J. E. Johnson, Jr., started a controversy.

Further, that cast iron containing oxygen has more combined carbon than that without it is the very contention made by those who oppose Johnson's theory that oxygen is directly the cause of higher strength, chilling power and hardness. Foundry metallurgists have always agreed with this after they became familiar with the oxygen situation. They hold that not the oxygen is the cause of strength, etc., but the effect oxygen has on raising the freezing point and thereby increasing the

combined carbon by quicker setting of the metal brings this about. There results a deeper chill, greater hardness and higher strength.

I further hold that these seemingly beneficial effects are heavily discounted by evils incident to the presence of oxygen, in the way of pin holes, gas pockets, excessive interior shrinkage, cracks resulting from inability of the metal to stretch in setting, etc., which more than counterbalance the apparent gain. It is dangerous to fool with oxygen in cast iron, as the concealed defects of a casting may come out in a most disagreeable fashion at a critical time.

Mr. Stork says that manganese added to metal high in oxygen takes up some of it and thereby reduces chilling power and strength. True, but this is because of a reduction in combined carbon. Outerbridge effected this many years ago by adding ferromanganese to his cast-iron car-wheel mixtures. He did not then know about the oxygen end of it, but overcame the difficulties this element caused and thereby got more reliable wheels.

Mr. Stork's statement that steel, when added to cupola mixtures, is oxidized and it is this oxygen content that gives such a mixture its "superior physical properties," I consider incorrect. It is the reduction in the total carbon brought about by steel additions that give the good qualities found in the resulting castings, the bad ones are attributable to the oxygen.

It is well known that the surface portion of a casting has a maximum of combined carbon, whereas the center of the heaviest section has the minimum. Graphite forms the complement of these figures, the sum of the two in each case, or total carbon, being practically uniform in amount throughout the casting. Unless, therefore, the sample for analysis for combined carbon and graphite is taken by a planer tool at an exactly specified depth below the skin, the results are not only incorrect but may be actually misleading. What value, then, have deductions made from combined carbon results except in a very general way? To be useful they should show a wider divergence than in the paper in question. That there is something amiss with some of the figures given is shown by adding up the combined carbon and graphite figures in Series A of the table, page 498. The metal poured hot has 3.25 total carbon, and after boiling the same iron has 3.60, which needs explanation. It is better in Series B.

The more extensive table, page 500, loses some of its value because of the unreliability of combined carbon determinations. The silicons can be grouped into three classes, by running a line between 1.18 and 1.36; and a line between 1.32 and 1.43. Mr. Stork admits that the results collected are from different days, and while they do bear out his contention that the chill decreased as the manganese went up, to the foundry metallurgist they must appear specially selected from a mass

of material not by any means bearing out the point in detail. For instance, below 0.50 manganese it matters little for results whether this element is 10 or 20 points apart. The sulfur is not given, and this is highly important in connection with manganese and chill results. In fact, picking out of the table three sets of figures, the following will be of interest:

Silicon	Manganese	Combined Carbon	Chill, Inch
1.18	0.29	1.47	0.80
1.35	0.35	1.32	0.75
1.43	0.34	1.49	0.75

For practical purposes, so far as these figures go, the chill is identical, the combined carbon is doubtful anyhow, and the manganese is about the same, yet the silicon, which can be determined with extreme accuracy, differs considerably. In regular practice there should have been a sufficient difference in chilling depth to correspond with the variation in silicon shown. All of which indicates that the table offers no particular proof on the effect of oxygen on cast iron.

In giving data on the slag situation in his melting practice, Mr. Stork, as already stated, only emphasizes the fact that this must have been defective in regard to the position of the bed, for when enough steel is burned to obviate the use of limestone in fairly long heats something is surely off.

On the whole, I am glad to see the paper published, as it affords a further opportunity to clarify a situation which most foundry metallurgists supposed had settled itself by going to sleep long ago.

Geo. F. Comstock,* Niagara Falls, N. Y. (written discussion†).—A study of this paper raises a question on which it is hoped Mr. Stork will throw more light; that is, why does an oxygen content generated on poor melting conditions have a deleterious effect on the properties of the iron, while an oxygen content generated by the oxidation of steel scrap in the cupola is beneficial to the strength and chilling ability? In each case the oxygen is absorbed at the same time and place in the process of melting and it is by no means clear why the superiority of semisteel should be assigned solely to its oxygen absorption, when oxygen absorbed without the steel addition is admitted to be injurious. It is not apparent why oxygen caused to be absorbed by the steel addition should not produce the effects of burning, sluggishness, slag inclusions, gas occlusions, etc., the same as oxygen absorbed by the iron in the cupola for any other reason, and it would seem that other factors to which he has not given sufficient emphasis must enter into the author's results. To an impartial reader of this paper, who was most favorably inclined toward J. E. Johnson, Jr.'s theory of the improvement of cast iron by oxygen, the discrepancy in the

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author's reasoning or statements is difficult to accept, and it is hoped that an explanation will appear in the final discussion.

R. S. MacPherran,* Milwaukee, Wis. (written discussion†).—Mr. Stork's paper is very interesting, but I am unable to agree with some of his conclusions. He argues very strongly for the beneficial effect of oxygen, but gives no actual determinations of this element to support his views. He states that the oxygen resulting from a high bed, heavy charges, low coke ratio, etc. must not be confused with the oxygen under normal conditions. I will be glad to know in what respect these oxygens differ or in what state they occur. Is it his opinion that the beneficial type is present as a dissolved gas and the other as metallic oxide?

I will not discuss the first five of his six points, as I have not seen enough trustworthy determinations of oxygen in cast iron to form any conclusions as to its effect. I have never known the addition of manganese, however, to reduce the strength of cast iron.

Under the sixth point, I would say that the addition of steel lowers the total carbon content, and by this action will increase the strength independently of oxygen. This reduction of carbon operates also to raise the melting point. It is interesting to note here that the author, in order to get proper fluidity for his cylinders, was obliged to reduce his oxygen by raising the manganese in the charge.

J. E. Johnson, Jr., in his paper on the influence of oxygen in cast iron,[4] refers to the determination of sulfur in his "wild iron" and says that in these irons he found the volatile sulfur low in comparison with the total. It would be interesting to know whether Mr. Stork made any experiments along this line.

The determination of oxygen will undoubtedly give interesting data and assist in the better understanding of both steel and iron. This determination, however, is difficult, and requires a high degree of technical skill.

A recent bulletin of the Bureau of Standards,[5] brings out this fact and rather puts in question many of the determinations previously made by this method. To quote from page 32 of this paper: "The Ledebur method requires extraordinary precautions to obtain reliable results. The errors we have described undoubtedly affect in greater or less degree nearly all results by this method that have been described in the literature, and if these are approximately correct, it is because of compensating errors."

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* Chief Chemist, Allis-Chalmers Mfg. Co. † Received Oct. 24, 1919.

[4] Influence on Quality of Cast Iron Exerted by Oxygen, Nitrogen and Some Other Elements. *Trans.* (1914) **50**, 344.

[5] J. R. Cain and Earl Pettijohn: A Critical Study of the Ledebur Method for Determining Oxygen in Steel. Bureau of Standards *Tech. Paper* 118 (1919).

W. L. STORK (author's reply to discussion*).—The criticism of the melting practice is unwarranted as the cupola practice was unequivocally according to the theories and practice of Doctor Moldenke, which were always found to give the desired results. When heats are run from 10 to 12 hr. without the sign of a flame at the charging door, the author considers the melting practice beyond criticism. As to the oxidation of carbon dioxide, the resulting gas will be carbon monoxide, which is combustible and would ignite at the charging door. Furthermore, a silicon content of 1.30 to 1.45 can hardly be called "remarkably low" and would not "throw the entire melting practice into that of making malleable." The 1.18 silicon heats were run only during the experimental period. The term oxidation is purely relative; all cupola melting is an oxidizing process, and it was one of the purposes of the paper and the contention of the author that the higher the steel content of the charge, the greater is the oxidation.

The author referred to the deoxidizing of an over-oxidized heat (burned iron) and not to the "deoxidizing of a heat after the steel has been oxidized," as stated by Doctor Moldenke. The author throughout his paper showed that the deoxidation of heat high in steel was possible and practical even with manganese itself. On page 499, the writer states that "the iron must be deoxidized with manganese." He is of the opinion, but regrets that he cannot locate the reference, that Doctor Moldenke advocates a high manganese content in mixtures carrying steel.

As to the difference between oxygen taken up by melting steel under proper melting conditions and that by ordinary mixtures under improper melting conditions, it is simply a matter of degree in that one has not gone as far as the other. Doctor Moldenke's remark on the boiling ladle that "if continued too long it is bad" proves that oxygen may have a good or a bad effect on the properties of cast iron depending on the extent of oxidation.

Referring to the question of oxygen increasing the strength of cast iron due to its increasing the combined carbon and not due to the oxygen itself is only a quibble. It is a case of the effect or the result. Likewise manganese removes oxygen, decreases its combined carbon, and thereby reduces chilling power and strength. The writer begs to add that his combined determinations were from the sample taken while the milling tool was cutting off the 4-in. riser from the impact bar and hence are reliable.

Doctor Moldenke's criticism of the table on page 500 sounds like an accusation that the results were garbled or padded; the author's conclusion, p. 501, "these results are not for the same ladles of iron but are from different days and are the general trend of conditions and results,"

* Received, Jan. 21, 1920.

is proof of his sincerity and purpose. The data were selected on a basis of like silicon contents. That these are results which do not bear out the law is to be expected but is not to be used as an argument against the theory. Every scientific man knows that almost any series of results when plotted or tabulated gives a number of cases that do not bear out the law yet it is permissible to plot them and to draw a mean curve through all the points. The very fact that the late J. E. Johnson, Jr., could start such a controversy by his theory is proof that the situation had not gone to sleep long ago.

As to whether oxidation of the cupola charge has a beneficial or deleterious effect on the properties of cast iron is a matter of degree, as has been admitted by Doctor Moldenke (see p. 503) and explained by the author. Oxidation of steel scrap melted under poor conditions can and will have the same bad effects on the iron as the oxidation of ordinary mixtures under poor melting conditions, whereas oxidized steel scrap melted under good conditions produces entirely different results. The author compared poor melting conditions with good melting conditions, hence the supposed discrepancy in reasoning.

The author made no actual oxygen determinations but used the Johnson theory of oxygen in cast iron and applied it to daily practice. The fact that manganese will reduce the combined carbon of a gray-iron mixture carrying oxygen is the reason for its reducing the strength. The addition of steel scrap to a cupola charge will not always lower the total carbon content. Under poor melting conditions, the carbon will not be reduced materially, as explained by Doctor Moldenke.[6] For instance, the addition of 40 to 50 per cent. steel scrap should bring the total carbon down to 2.85, as stated by Doctor Moldenke[7] and experienced by the author and others. Yet many of the shell manufacturers found it almost impossible to reduce the carbon content below 3.00 per cent. It is a matter of melting conditions whether there is a lowering of the total carbon or how low it shall be.

[6] "Principles of Iron Founding" 201. [7] *Op. cit.*, 199.

Graphitization of White Cast Iron upon Annealing*

BY PAUL D. MERICA,† PH. D., AND LOUIS J. GUREVICH,‡ B. CH. E., WASHINGTON, D. C.

(Chicago Meeting, September, 1919)

IN connection with other investigations on the properties and characteristics of chilled-iron car wheels, the question as to the best range of annealing temperatures was raised. Chilled-iron wheels are cast with a chill against the tread and the inside of the flange, the remainder being in sand. The composition of the metal is so chosen that under these conditions the tread and inside of the flange will show white iron to a depth of from 5/8 to 7/8 in. (15 to 22 mm.), the remainder of the wheel becoming graphitized or gray. In order to relieve the stresses set up during cooling under such drastic conditions, the wheels are stripped from the mold, while still red hot, piled in a soaking pit, and allowed to cool very slowly from their temperature at stripping.

Obviously the most suitable temperature for this annealing is the highest at which no formation of graphite occurs within the white tread and at which the wheels can be stripped from the molds. Inasmuch as no direct determinations have been made of the temperatures at which the formation of graphite takes place in white iron of compositions used in car wheels upon annealing, it was considered worth while to determine these temperatures as a means of establishing the maximum temperatures at which the annealing of the wheels may be carried out. In the course of this work, some incidental observations were made which are of interest in connection with the theory of graphitization in white iron.

The literature on the subject of the formation of graphite both during cooling at casting and upon annealing is very extensive and can be best studied by reference to a comprehensive discussion of the subject by Hatfield.[1] Perhaps the most important contributions of previous investigation are the following:

1. Whether a given mass of molten iron becomes white, mottled, or gray upon casting, depends on its composition and its rate of cooling through the temperature range of solidification and immediately below.

2. The more rapid the rate of cooling, the less is the extent of graphitization.

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* U. S. Bureau of Standards *Tech. Paper* 129 (1919).

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‡ Assistant Physicist, U. S. Bureau of Standards.

[1] W. H. Hatfield: "Cast Iron in the Light of Recent Research," Chas. Griffin & Co., 1918.

3. Graphitization increases with increase of carbon or silicon content; it decreases with increasing sulfur content. Manganese plays a double role—by combining with the sulfur it first neutralizes the specific effect of the sulfur; an excess of manganese restrains graphitization.

4. Upon annealing irons that have solidified white, the combined carbon decomposes into graphite or temper carbon at temperatures from 700° to 1100° C., depending on the period of annealing and the composition of the metal. Those elements that promote graphitization upon casting and solidification promote it also during annealing by lowering the temperature range within which decomposition of the cementite is possible.

5. After nuclei of graphite have once formed, further graphitization proceeds more readily and at lower temperatures than in iron of the same composition in which no nuclei are present.

Materials and Annealing

All of the samples used in these experiments were kindly furnished by the Griffin Wheel Co. through the courtesy of Mr. F. K. Vial. Their chemical compositions are given in Table 1. The samples of the I and the II series were cast in green sand as ¾-in. (19-mm.) cubes; those of the CB and CF series were cast in an iron-chill mold as cylinders of 1-in. (25-mm.) diameter; the samples of series 5 were cast in dry sand as ¾-in. cubes. Series I, II, and 5 were poured from the same heat of metal.

Table 1.—*Chemical Compositions of Materials Used*

Heat or Series No.	Character	Total Carbon, Per Cent.	Graphite, Per Cent.	Combined Carbon, Per Cent.	Silicon, Per Cent.	Sulfur, Per Cent.	Phosphorus, Per Cent.	Manganese, Per Cent.
I	White	3.67	0.04		0.57	0.178	0.34	0.53
II	White	3.68	0.09		0.55	0.163	0.33	0.54
CB	White	3.78	0.02		0.57	0.100		0.63
CF	White	3.90	0.02		0.64	0.204		0.63
5	Gray	3.66	2.80	0.86	0.63	0.185	0.31	0.58

The samples were annealed in an electric furnace of the laboratory type at temperatures and for periods given in Tables 2, 3, and 4. In the earlier part of the work, the specimens were packed in asbestos fiber and annealed in air. It was found, however, that too much decarburization took place at the higher temperatures, so the samples were thereafter annealed in a current of dry nitrogen.

In order to keep the temperature, as measured by a thermocouple attached to the specimen, constant during the period of anneal, the current for the furnace was, in most cases, supplied by a generator with a Tirrell regulator giving a constant voltage to within approximately ±

1 per cent. In this manner it was possible to maintain the temperature of the specimen constant during the period of anneal to within approximately ± 5° C. The authors wish to express their appreciation of the aid given by R. W. Woodward in assembling the furnace for annealing and the current regulating apparatus. All of the specimens, with the exception of those marked air-cooled or quenched, in the table, were cooled in the furnace to room temperature, a temperature recorder indicating the rate of cooling. For the three furnaces used, these cooling rates are given below:

Furnace	From 800° to 700° C.	From 700° to 300° C.
Platinum-wound	19 min.	143 min.
Nichrome	8 min.	70 min.
Nichrome-tube	26 min.	160 min.

The furnace used for each specimen is stated in the tables.

Table 2.—*Annealing of White Cast Iron, Series I*

Specimen No.	Temperature, Degrees C.	Graphite, Per Cent.	Combined Carbon, Per Cent.	Total Carbon, Per Cent.
Annealed 6 hr. in platinum-wound furnace; cooled in furnace				
Io	As received	0.05	(3.61)	(3.64)
IC	700	0.05	(3.61)	(3.64)
ID	800	0.09	(3.55)	(3.64)
IM	820	0.12	(3.52)	(3.64)
IN	840	0.40	(3.24)	(3.64)
IO	860	1.31	(2.33)	(3.64)
IP	880	1.92	(1.72)	(3.64)
IE	900	2.74	(0.90)	(3.64)
IS	950[a]	2.48	(1.16)	(3.64)
IT	950	2.85	0.74	3.59
IF	1000	3.25	0.26	3.51
IIF	1000	3.27	0.34	3.61
IV	1000[b]	3.44	0.20	3.64
IG	1100	2.56	0.70	3.26
IU	1100[b]	3.12	0.52	3.64
Annealed 48 hr. in nichrome furnace in nitrogen; cooled in furnace				
IH	700	0.20		
Ii	750	0.35		
IJ	775	1.60		
IL	800	2.90		
IQ	900	3.30		
	850	3.21	0.36	3.57
IK	800[c]	0.74		

[a] Quenched in water. [b] Annealed in nitrogen.
[c] Annealed 24 hr. and cooled in furnace.

TABLE 3.—*Annealing of Gray Cast Iron; Series 5**

Specimen No.	Temperature, Degrees C.	Graphite, Per Cent.	Combined Carbon, Per Cent.	Total Carbon, Per Cent.
5	As received	2.80	0.86	3.66
5F1	600	2.87	(0.83)	
5E1	700	3.08	(0.62)	
5E2	700	3.02	0.68	3.70
5D1	800	3.06	(0.64)	
5D2	800	3.08	(0.62)	
5C1	900	3.28	(0.42)	
5C2	900			
5B	1000[a]	3.44	(0.26)	
5A	1100[a]	2.98	0.66	3.64

[a] Annealed in nitrogen.

* Annealed 6 hr. in platinum-wound furnace; cooled in furnace.

TABLE 4.—*Annealing of White Cast Iron; Series CB and CF*†

Specimen No.	Temperature, Degrees C.	Graphite, Per Cent.	Specimen No.	Temperature, Degrees C.	Graphite, Per Cent.
CB	As received	0.02	CF	As received	0.02
CB5	720	0.02	CF1	720	0.03
CB8	780	0.04	CF8	820	0.13
CB1	820	0.83	CF9	840	0.94
CB6	845	2.59	CF12[a]	850	2.03
CB11[a]	850	2.80	CF6	855	2.21
CB12[a]	870	3.38	CF13[a]	870	3.33
CB7	890	3.50	CF7	890	3.45
CB13[a]	900	3.51	CF14[a]	900	3.64
CB2	920	3.45	CF3	920	3.48
CB3	1020	3.42	CF4	1020	3.72
CB4	1120	3.25	CF5	1120	3.47

[a] Annealed in nichrome tube furnace.

† Annealed 6 hr. in nichrome furnace in nitrogen, and cooled in furnace.

The graphite content of the annealed specimens was determined by a modification of the usual method developed by Dr. E. Schramm, who made many of the analyses reported below. The method is to dissolve the sample in nitric and hydrofluoric acids and weigh the insoluble residue after washing. As a check, the residue is ignited and weighed again. The total carbon content is determined by combustion and the combined carbon calculated by difference.

The results of all of the determinations are given in Tables 2, 3, and 4. The figures in parentheses represent values obtained on specimens from the same heat but not on the sample opposite. The results are also summarized in Figs. 1 and 2.

Discussion of Results

It will be observed that for the composition I, the annealing or graphitization range of temperature was much lower for the 48-hr. annealing period (700° to 800° C.) than for the 6-hr. period (800° to 950° C.); the range is also narrower. These facts are in accord with experience in the annealing of white iron.

The effect of variation of chemical composition within the limits of composition studied, upon the graphitization range of temperature, was not great. The effect of difference in total carbon content was ap-

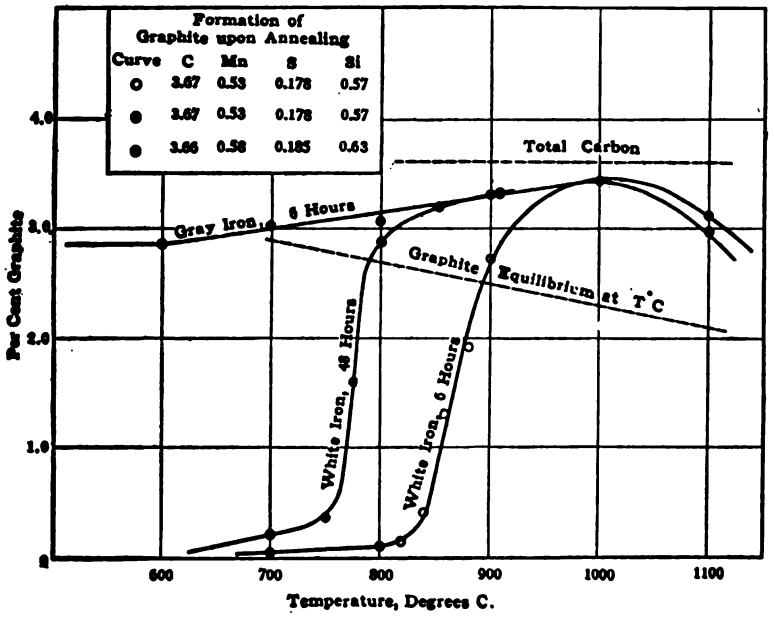

Fig. 1.—Formation of graphite on annealing cast iron. Compositions of iron are as given.

parently greater than that of difference in sulfur content. Thus, both series, CB and CF, having higher total carbon content than I or II, showed a complete precipitation of temper carbon within a narrower temperature range than did I or II, although the temperature of incipient decomposition was very nearly the same in all cases for the 6-hr. annealing period. The effect of variation of sulfur content from 0.10 to 0.20 per cent. was overshadowed by a variation of total carbon content from 3.60 to 3.90 per cent. It is apparent that the manganese present in all of the specimens was sufficient to neutralize practically entirely the effect of the sulfur. Microscopic examination of the samples confirmed this view in that a large amount of manganese sulfide was visible. From

the standpoint of the commercial annealing of car wheels, the usual variations in chemical composition will not apparently be of any significance. The effect of period of annealing is, however, quite marked, as is shown by Fig. 1.

The maximum allowable temperature for "pitting" wheels is 730° C.; above this temperature there is danger of some softening of the tread due to graphitization during annealing.

It is noted, also, in entire accord with previous experience, that during a 6-hr. anneal the gray iron 5 of the same heat or composition as I or

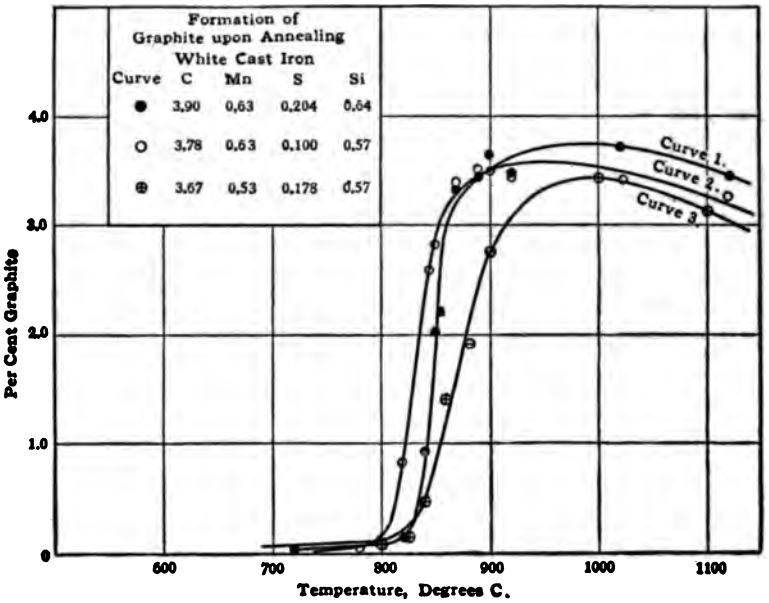

FIG. 2.—FORMATION OF GRAPHITE ON ANNEALING WHITE CAST IRON. ALL SPECIMENS WERE ANNEALED FOR 6 HR. COMPOSITIONS ARE AS GIVEN.

II, white, begins to graphitize further at temperatures lower than those at which there is initial graphitization of the white iron.

Attention is invited to the curious fact indicated in the figures, that the specimens annealed at 1100° C. and cooled in the furnace showed less graphite and consequently more combined carbon than those annealed for the same period at 1000° C. and cooled at the same rate. This fact was established by several determinations and is confirmed by microscopic examination of the annealed samples. Figs. 3, 4, 5, and 6 show the microstructure of samples CB3, CB4, IF, and IU, respectively, after annealing in nitrogen. Specimens CB4 and IU, annealed at 1100°, have more pearlite than CB3 and IF, annealed at 1000° C.

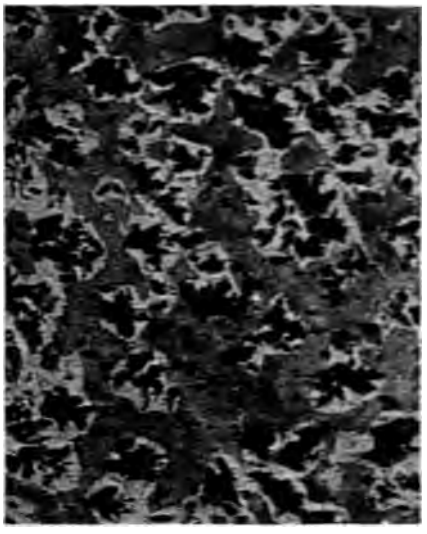

FIG. 3.—SPECIMEN CB3; ANNEALED 6 HR. AT 1020° C. × 100.

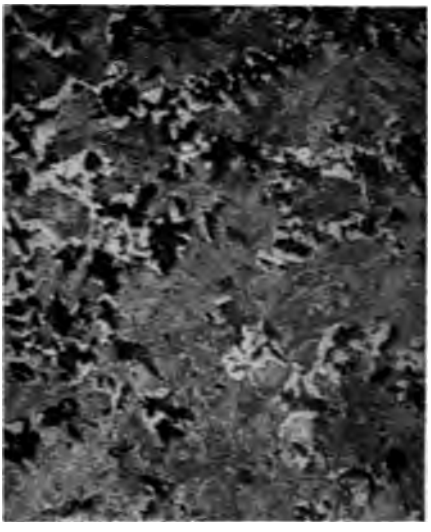

FIG. 4.—SPECIMEN CB4; ANNEALED 6 HR. AT 1120° C. × 100.

FIG. 5.—SPECIMEN 1F; ANNEALED 6 HR. AT 1000° C. × 100.

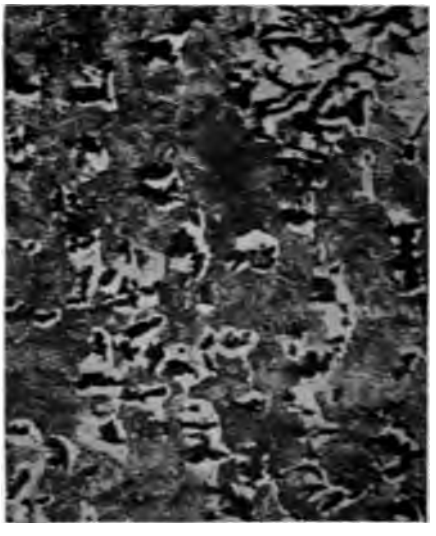

FIG. 6.—SPECIMEN 1U; ANNEALED 6 HR. AT 1110° C. × 100.

FIGS. 3 TO 6.—MICROSTRUCTURE OF WHITE CAST IRON ANNEALED IN NITROGEN.

In Fig. 7 is reproduced a part of the equilibrium diagram of the iron-carbon alloys. The line *ES* represents equilibrium between cementite (Fe_3C) and solid solution; the line *BJ*, which has never been accurately determined, represents the stable equilibrium between graphite and solid solution.

Upon annealing white iron at increasingly high temperatures, the free cementite is first decomposed into graphite and the total graphite content is thus increased. At a sufficiently high temperature, about 1000° C. in the instance described, all the free cementite has been decomposed

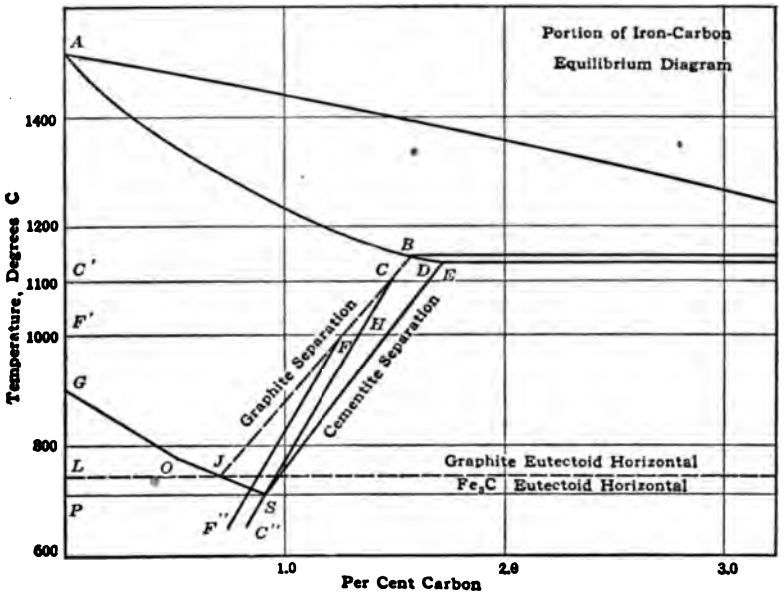

FIG. 7.—PORTION OF IRON-CARBON EQUILIBRIUM DIAGRAM.

and equilibrium obtains between graphite and solid solution (line *BJ*). Thus, at 1100° there is an amount of carbon in solid solution equal to *C'C*, at 1000°, an amount equal to *F'F*, which is less than *C'C*. The graphite content of specimens that have attained equilibrium at such high temperatures and thereupon been quenched, would be found to correspond to these amounts, as they were in one experiment.

Upon cooling, at a more moderate rate, however, there is sufficient time for crystallization of the graphite upon the nuclei already present, to correspond more closely to equilibrium solubility. If this precipitation were complete at each moment and equilibrium obtained, the amount of combined carbon in a sample annealed at 1100° and cooled to 1000° would be exactly equal to that in a sample annealed at 1000°; there could be, consequently, no difference in the combined carbon content of the two

samples, after cooling at the same rate from this temperature to room temperature. As a matter of fact, there is a marked difference between the two carbon contents. How is this to be explained?

It is often held that graphitization during cooling from 1100° to 700° C. is due to the formation of the temper carbon from cementite, which has previously precipitated along the line *ES*, and that graphite itself can not form directly from the solid solution. Such a theory of the mechanism of the formation of graphite is not capable of explaining the phenomenon described, since no free cementite is found after cooling from the high temperatures. The decomposition of the cementite upon its appearance must therefore have proceeded promptly, and the amount of pearlite after complete cooling would necessarily correspond to the amount present in the pearlite eutectoid.

The graphite must separate directly from the solid solution, but there is apparently a lag in its separation at these rates of cooling. Thus, the amount of combined carbon during cooling at any instant must have corresponded to some such values as lie on the line *CC''*, upon cooling from 1100°, and upon the line *FF''*, upon cooling from 1000°. At the temperature of intersection of these lines with the pearlite eutectoid horizontal, all of the cementite still remaining in solid solution precipitates as pearlite. The difference in the amount of combined carbon left after cooling from 1100° and from 1000° C. is thus explained by the lag in the precipitation of the graphite along its equilibrium curve.

Since the amount of combined carbon finally formed as pearlite in this manner must be greater than the amount of combined carbon at the graphite eutectoid, inasmuch as the lines *FF''* and *CC''* lie to the right of the line *BJ* of graphite equilibrium, the results given would seem to indicate that the graphite eutectoid lies at a smaller value of carbon content than has been previously supposed. At least this is true unless there is either a marked formation of graphite eutectoid at these rates of cooling, or a decomposition of pearlite into graphite, both rather unlikely suppositions but not impossible ones. The lowest value of combined carbon as pearlite in annealed and slowly cooled specimens found in these experiments was 0.20 per cent.; Guertler[2] gives 0.70 per cent. as the most likely amount of carbon in the graphite eutectoid.

The authors have not had an opportunity to continue the study of this phase of the matter, but it would appear that further investigation into the subject of graphitization upon annealing and cooling were needed. An exact determination of the course of the line *BJ* of graphite equilibrium and of the rate of precipitation of graphite at different temperatures from solid solution would be of much practical value as well as possess scientific interest.

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[2] W. Guertler: "Metallographie." Berlin, 1912. Borntraeger.

SUMMARY

1. The annealing or graphitization ranges of temperatures were determined for three different compositions used for car wheels. The temperature of initial precipitation of temper carbon for 6 hr. of annealing was not noticeably affected by variation of sulfur content from 0.10 to 0.20 per cent. or by variation of total carbon content from 3.60 to 3.90 per cent., although the effect of greater carbon content is to narrow the temperature range within which graphitization is complete.

2. The temperature of beginning precipitation of temper carbon was about 830° for the 6-hr. period of annealing, and about 730° C. for the 48-hr. period. The maximum allowable temperature, therefore, for the annealing, or "pitting," of car wheels is about 730° C.

3. After complete decomposition of all free cementite by annealing at from 1000° to 1100° C. and cooling at equal rates in a laboratory electric furnace, less graphite is found in a specimen cooled from 1100° than in one of the same composition cooled from 1000° C. This indicates that graphite separates directly from solid solution upon cooling, when its nuclei are already present.

4. The fact that only 0.20 per cent. of combined carbon was found in some specimens after annealing at high temperatures and cooling slowly in the furnace would indicate either that the graphite eutectoid lies at much lower values of carbon content than has been previously supposed, that there is at those rates of cooling a direct precipitation of graphite eutectoid, or that there is a formation of graphite from pearlite at temperatures directly below that of its formation.

DISCUSSION

H. A. SCHWARTZ,* Indianapolis, Ind. (written discussion†).—Referring to the very interesting paper on the graphitization of white cast iron, the writer desires to present certain facts and conclusions, entirely in harmony with those of the authors but which may be not without interest as furnishing additional evidence upon the same subject.

The graphitization of white cast iron, being the basis of malleable "annealing," has naturally been of very great interest to the National Malleable Castings Co., and has been the subject of investigation for some 12 or 15 years in the writer's laboratory. Our early work was somewhat erratic, but was given better direction by the published results of Doctor Storey, and after that date progressed easily and rapidly to what we regarded as a satisfactory conclusion. The latter and more successful part of our work consisted of determining, not the amount of graphite formed after exposure to a given temperature for a given time,

* National Malleable Castings Co. † Received Sept. 4, 1919.

but the time and concentration of combined carbon when equilibrium was attained at a given temperature. In general, a series of samples of uniform material of known composition would be held at a given constant temperature for various times, quenched and analyzed for total and graphitic carbon. The microscopic changes were also followed carefully. The exigencies of the available time and equipment did not permit of either the precision or completeness desirable if the results are to be applied to the construction of an equilibrium diagram for the stable system Fe-C, but the following conclusions are deemed to be justified by the observed facts.

Graphitization can be initiated at any temperature at or above a definite critical point. This point is very near the A_1 point of the system Fe_3C; it may possibly not be identical with it, but may be found at a slightly lower temperature. A number of observers have reported the initiation of graphitization at temperatures slightly but decidedly below A_1. The writer has confirmed this observation, but there remains some doubt as to whether the divergence is due to errors of observation or is real.

The time to attain equilibrium, in a given alloy, is inversely as the temperature above the critical point, but varies with the chemical composition of the material and with its previous thermal history. Equilibrium when attained should correspond to *EJ* of the authors' paper. This line joins $CO = T$ = critical point; with the point E approximately C = 1.7 per cent., T = 1130° C., and the line is approximately, though not necessarily exactly, straight. We are more certain of its location toward the lower end than at temperatures above 900° C.

Heating at any given temperature indefinitely produces no increase in graphite content at the expense of combined carbon after the combined carbon has dropped to a concentration corresponding to equilibrium at the temperature used. The microstructure then corresponds to a homogeneous mix crystal matrix through which is scattered free carbon. On cooling such material to a lower temperature, the structure and composition may be made to conform to that in equilibrium with free carbon at each succeeding lower temperature by sufficiently retarding the rate of cooling. In general, cooling must vary with the elevation of temperature above the critical point, being, in theory, infinitely slow when A_1 is nearly reached. Even at fairly high temperatures rates of 5° C. an hour are rather fast. It is, therefore, very difficult to actually attain the equilibrium condition at relatively low temperatures. At temperatures at or below the critical point the equilibrium condition is that in which only ferrite and temper carbon exist, as in malleable cast iron; this state, however, is not reached unless cooling is slow. Cooling to A_1 too rapidly to permit the alloy to reach equilibrium at each successive temperature during cooling permits the alloy to reach A_1 while still containing considerable amounts of carbon in the "com-

bined" state. At A_1 such alloys allow the mix crystal to be converted into pearlite and form structures exactly similar to the authors' Figs. 3 to 6, temper carbon surrounded by ferrite with a background of pearlite.

The very slow initiation and progress of graphitization at low temperatures is clearly shown by the comparison of the 700° points in the 6-hr. and 48-hr. branches of the curve in Fig. 1. The more rapid progress is shown in the 800° points. At 900° and 1000°, equilibrium being nearly approached even in 6 hr., the increased graphitization with additional time is not marked.

The reason for the apparent anomaly referred to in the last paragraph of page 516 is, in the writer's opinion, capable of explanation on the following assumptions, which are identical with the authors' explanation in reasoning, but differ in the wording. The time of heating, being the same for 1000° and 1100° and sufficient to attain equilibrium at 1000°, was also necessarily sufficient to leave the metal in equilibrium at 1100°. The metal at 1100° contains more combined carbon than that at 1000°, since carbon is more soluble at the higher temperature, indicated by the direction of *BJ*. Cooling was more rapid than is consistent with the maintenance of equilibrium with descending temperature, therefore the sample heated at 1100° when it reaches 1000° has more combined carbon than the sample held at 1000° until equilibrium was attained. Both samples cooling from 1000° at the same rather rapid rate, the sample which at 1000° had the highest combined carbon will also arrive at A_1 with the highest carbon and will therefore contain more pearlite than its mate. The same phenomenon is not noted at lower pairs of temperature, since in those cases initial equilibrium was not attained.

The authors' conclusion that the eutectoid point for the system Fe-C is much lower than Guertler's figure, 0.70 per cent., cannot be successfully attacked. Aside from research experiments, there is the evidence of the entire tonnage of malleable castings which represent the end point of the graphitizing reaction and in which the combined carbon is seldom above 0.10 per cent. Data such as Guertler's are frequently based on the apparent termination of the graphitizing reaction during the freezing of gray iron coupled with an inadequate appreciation of the fact that graphitization occurs in two distinct stages.

The mechanism by which the graphitization of white cast iron is accomplished seems not to be clearly understood as yet. There seems to be a general acceptance of the fact that the presence of primary cementite is a requisite to the initiation of graphitization. This conclusion is pretty well supported by the fact that a carbon content much below 2 per cent. either prevents or retards graphitization very materially. Furthermore, it has been observed that alloys of such carbon content as to be non-eutectiferous but hypereutectoid, *i.e.*, approximately between

0.9 per cent. carbon and 1.7 per cent. carbon, may graphitize better at low temperatures, when they fall to the right of *SE*, than at higher temperature, when they are to the left of that line and hence contain no free cementite. This thought, however, would require that graphitization should terminate at any given temperature at a (combined) carbon concentration corresponding to the intersection of the line *ES* with that temperature. Doctor Merica's observations indicate that this is not the case, graphitization terminates at points on the line *BJ* to the left of *ES* as graphitization continues after the destruction of cementite is complete. The fact that graphitization proceeds at measurable rates under conditions at which it cannot be readily initiated is explained by the effect of the presence of carbon nuclei. The fact that *BJ* has been determined in our laboratories as intersecting the line *CO* very close to the A_1 point and that it is approximately straight indicates apparently conclusively that the reaction is completed by the graphitization of the mix crystal.

The straightness and location of *BJ* preclude the thought that graphitization extends only to the destruction of free cementite and that graphitization if complete terminates necessarily by the graphitization of the cementite of pearlite below A_1. This supposition would give as the locus of *BJ* the line *ES* followed by a line joining *S* with the critical point of the stable system, *CO*, temperature slightly under A_1. There seems some evidence, however, in the data of the writer confirmed by observations by Touceda, that upon reheating the completely graphitized product no recombination of free carbon occurs till Ar_1 is passed and then the concentration of combined carbon is of the eutectoid ratio. The equilibrium in that case, however, is then no longer that of the stable system, but of the metastable system Fe_3C.

The writer is strongly convinced that the line *BJ* now regarded as not well established in location by most writers should be provisionally drawn as a straight line to intersect *CO* at about 680° C. A very accurate determination of its location would be extremely interesting and might slightly alter its form and location.

In the freezing of gray iron it is true that graphitization ends at concentrations of combined carbon of the general order of magnitude of the carbon in the eutectoid (pearlite). This seems to be due to the fact that the time at higher temperatures has been such as to permit the complete destruction of cementite but the cooling has been too rapid to permit of much graphitization in the mix crystal.

A Volute Aging Break

BY HENRY M. HOWE AND EDWARD C. GROESBECK, BEDFORD HILLS, N. Y.

(New York Meeting, February, 1919)

(A CONTRIBUTION FROM GREEN PEACE LABORATORY)

FIG. 1 shows a volute aging break which developed spontaneously in a hardened and tempered steel helmet between 19 and 38 days after it had been tested ballistically.

A similar break, shown in Fig. 2, developed spontaneously in a similar helmet about 40 days after the ballistic testing. In this case the rupture occurred abruptly and with a loud report. The path of rupture in this second case, though much the less regular, is the more interesting of the

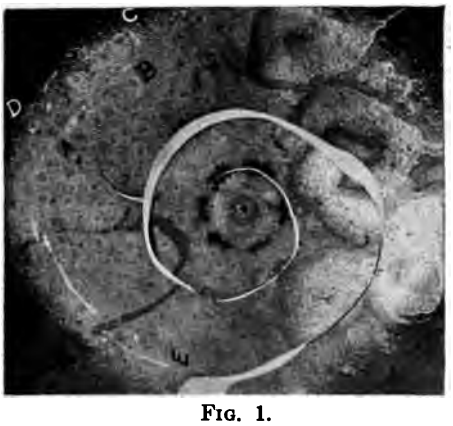

FIG. 1.

FIG. 2.

two, in that it is an approximation to two parallel volutes centered at the point of impact of the projectile, just below *D*. Several weeks later, after the photograph for Fig. 1 was taken, a similar doubling of the volute crack occurred in this helmet, starting from the end of the eyelash crack near *A* and spreading thence to the left contra-clockwise for about ½ in.

The negatives from which these pictures are made have not been touched up in any way, save writing the letters on them.

In both cases the trajectory of the bullet was normal to the initial surface at the point of impact, and in both the helmet passed the ballistic test without detectable cracking, under a velocity of impact not far below that which usually causes perforation. The initial surface itself was very nearly spherical throughout the region covered by the break.

In the early days of making armor-piercing shells, spontaneous and violent aging rupture was so common that the shells, after hardening, used to be stored for a considerable length of time in a room to which nobody was admitted.

Since the foregoing was written several additional volute breaks have been noticed, one at the opposite end of the helmet shown in Fig. 1, and one in another helmet. This last is at least as regular as that shown in Fig. 1.

These later cracks occurred about three months after the ballistic testing ended. About two weeks later the attention of one of us was arrested by a succession of faint cracklings proceeding from a pile of tested helmets.

Path of Rupture in Steel Fusion Welds*

BY S. W. MILLER,† ROCHESTER, N. Y.

(New York Meeting, February, 1919)

MOST of the steel welding done at the present time is in material containing not over 0.3 per cent. carbon, and the tests here described were in similar material. These tests are not as yet completed but it is hoped that more results and more reasons for the conclusions may be presented at the meeting in February.

From the time the author found the peculiar structure in electric welds referred to in a previous paper,[1] that is, needles or plates in the grains and similar material, in larger spots, at the grain boundaries, he has felt that these plates which may be iron nitride and probably contain some carbon, were at least partly responsible for the brittleness of electric welds. He has found but few traces of similar lines in oxyacetylene welds until the last 2 mo., he has since found large numbers under certain conditions. In the top of these welds, they are quite numerous. In the body of heavy oxyacetylene welds, they appear at times, but not nearly to the extent that they do in electric welds; the author has never seen them in the body of welds in material less than $\frac{3}{4}$ in. (19 mm.) thick. They are shown in Fig. 82. The reason they were not noticed before is undoubtedly due to the fact that in making tests of welded pieces the weld was always ground off level with the plate, and as this structure only appears in the top $\frac{1}{16}$ in. (1.5 mm.) of the weld, they were removed by the grinding or machining; also, welds $\frac{3}{4}$ in. thick had not been examined.

It did not appear, in making bending tests, that this structure had much influence on the strength of oxyacetylene welds; while electric welds are noticeably brittle. Further, oxyacetylene welds made with certain special materials were exceedingly brittle even after the tops of the welds were removed; this was true in cases where there was no sign of the plates whatever and where the welds were remarkably free from oxide and the other usual defects. It appeared, therefore, that this brittleness did not depend on any variable that had been noticed so far, and it was decided to find where the rupture occurred in various kinds of welds made

* Report of Research under the joint auspices of National Research Council and Emergency Fleet Corporation.

† Proprietor, Rochester Welding Works.

[1] *Trans.* (1918) **58**, 700.

with various materials, in order to determine, if possible, the cause of the brittleness that existed under such widely varying conditions.

A small testing machine was made, in which a specimen $\frac{3}{8}$ in. (9.5 mm.) wide and up to $\frac{1}{4}$ in. (6.3 mm.) thick could be bent, after being polished and etched, and examined under the microscope. The results have been quite satisfactory, although unexpected, and the author hopes that further tests will make matters still more clear.

Table 1 gives the information regarding the pieces referred to by the photographs. The figures given for carbon, etc., are approximate only for no analysis has been made. All these pieces were tested by bending except EWQ, EWN, CBE1, and BS1. The work as planned includes further tests of other pieces and of pieces that are heat treated in various ways. There are many difficulties in attempting to present, by means of photographs, results that are more easily seen under the microscope. Among them is the trouble of finding fields that can be photographed; also, most of the distortions are so out of focus that they cannot be photographed, although they can be readily seen with the eye. Again, it has been found that an exposure about four or five times as long as with surfaces that are etched in the usual way, and that are not distorted is

TABLE 1.—*Description of Test Pieces*

Mark	Original Material	Weld Material	Process	Thickness of Material Inch
CF2	Tank plate	Vanadium steel	Oxyacetylene	0.50
RA1	Tank plate	Roebling oxyacetylene welding wire	Oxyacetylene	0.50
RE2	Tank plate	Roebling electric welding wire	Oxyacetylene	0.50
106	Ship plate	0.25 per cent. carbon steel wire	Electric	0.50
40C	Tank plate	0.40 per cent. carbon steel wire	Oxyacetylene	0.50
4C4	Tank plate	0.40 per cent. carbon steel wire	Oxyacetylene	0.50
Q	Bar steel	Not known, covered electrode	Electric	0.25
A	Armco iron	Not welded		0.50
EWQ	Tank plate	Roebling electric welding wire	Electric	0.50
EWN	Tank plate	Roebling electric welding wire	Electric	0.50
CBE1	Tank plate	Roebling electric welding wire with 2 per per cent. aluminum	Carbon arc	0.50
BSI	Ship plate	Roebling oxyacetylene welding wire	Oxyacetylene	0.75

ARMCO IRON.

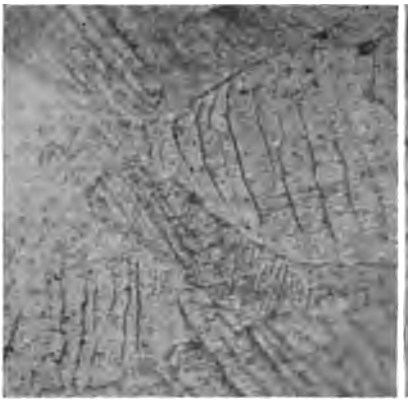

FIG. 1.

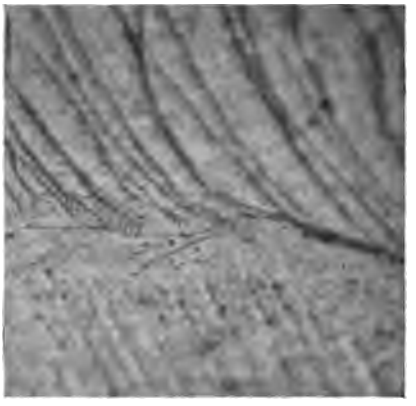

FIG. 2.

FIG. 1.—AFTER STRAIN SLIP BANDS APPEAR IN NEARLY EVERY GRAIN. × 430.
FIG. 2.—SLIP BANDS CURVE TOWARD GRAIN BOUNDARIES. × 430.

FIG. 3.

FIG. 4.

FIG. 3.—ANOTHER SET OF CURVED SLIP BANDS; VERY COMMON IN THIS MATERIAL. × 430.
FIG. 4.—SLIP BANDS INTEGRATING INTO GRAIN BOUNDARY RUPTURE. EDGES OF CRACK ROUNDED, NOT SHARP AS IN CF2, 106, ETC. × 430.

In Armco iron the grain boundaries are stronger than the slip planes. Because of higher tensile strength of CF2, 106, etc., their grain boundaries and slip planes are probably stronger than the slip planes of Armco iron, though they may be weaker than Armco iron grain boundaries.

SPECIMEN 40C.

FIG. 5.

FIG. 6.

FIG. 5.—USUAL STRUCTURE OF WELD. PEARLITE AND CEMENTITE AT GRAIN BOUNDARIES. × 430.

FIG. 6.—OXIDE FILMS AT GRAIN BOUNDARIES $\frac{1}{40,000}$ IN. THICK. × 430.

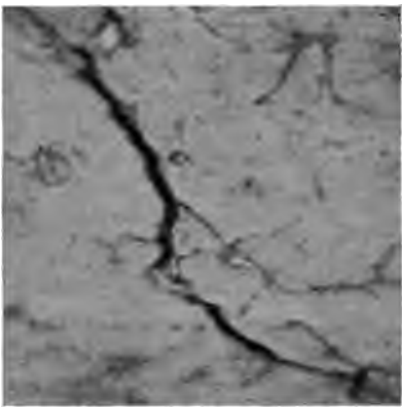

FIG. 7.—GRAIN BOUNDARY RUPTURE AT OXIDE FILMS OF FIG. 6. × 430.

SPECIMEN 106.

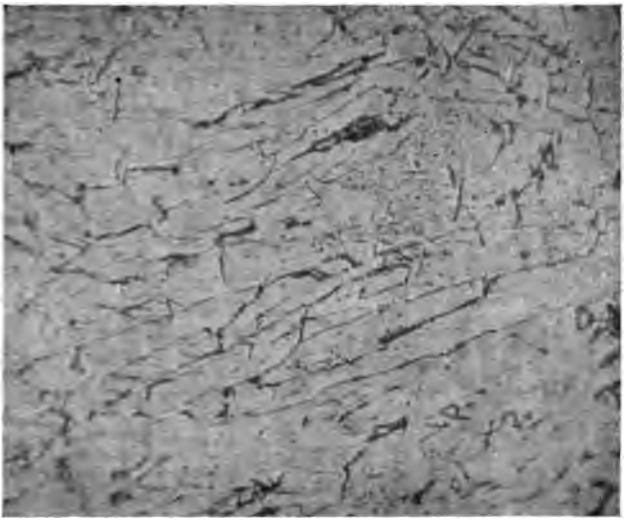

FIG. 8.—USUAL STRUCTURE OF COLUMNAR GRAINS, WITH LINES AND SPOTS AS USUAL IN ELECTRIC WELDS. × 430.

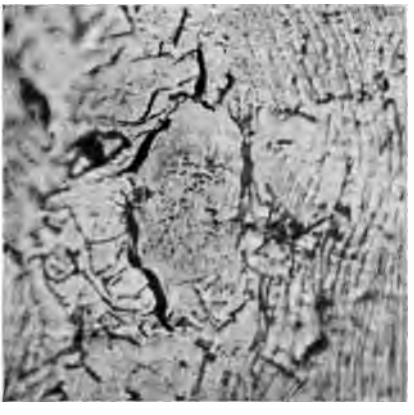

FIG. 9.

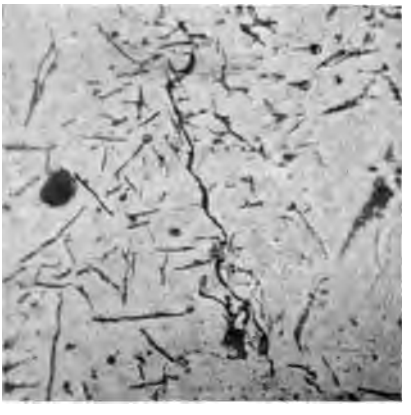

FIG. 10.

FIG. 9.—INTERGRANULAR RUPTURE, POSSIBLY DUE TO OXIDE FILMS. × 430.
FIG. 10.—INTERGRANULAR RUPTURE, NOT DUE TO OXIDE FILMS. × 430.

SPECIMEN 106.

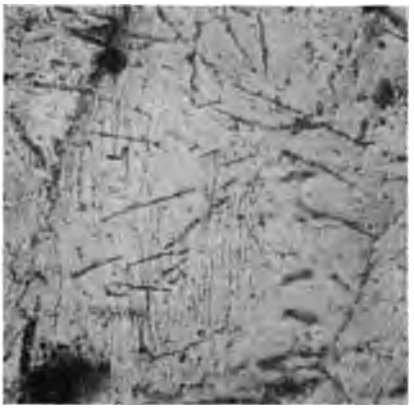

FIG. 11.

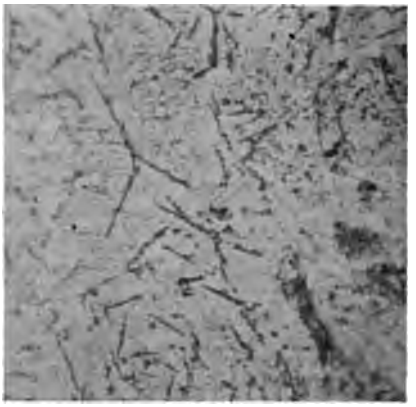

FIG. 12.

FIG. 11.—SLIP BANDS, INDEPENDENT OF DIRECTIONS OF LINES. × 430.
FIG. 12.—SLIP BANDS AT 45° TO DIRECTION OF STRAIN. × 430.

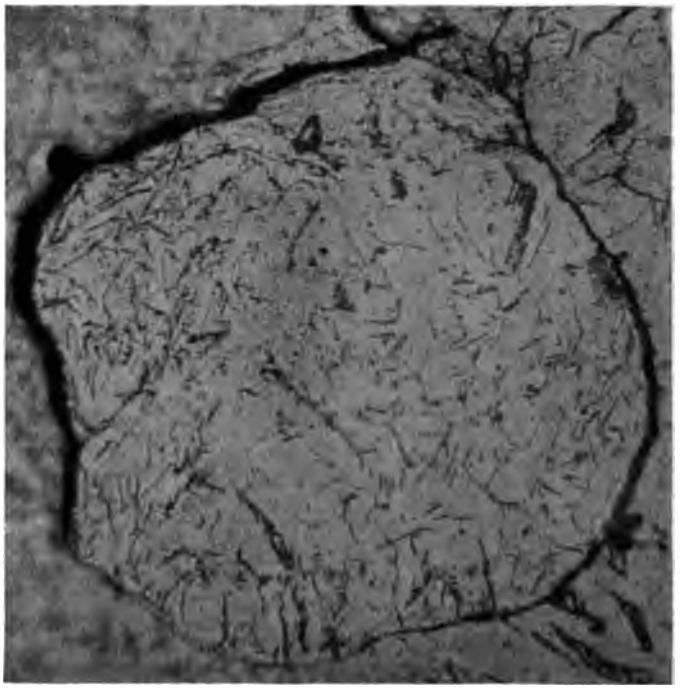

FIG. 13.—GLOBULE OF IRON SURROUNDED BY FILM OF OXIDE. STRAIN HAS OPENED THE CRACK. × 430.

SPECIMEN 106.

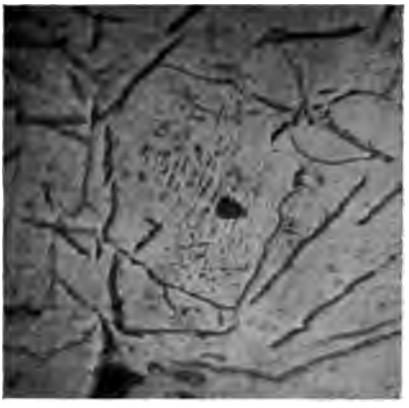

FIG. 14.

FIG. 15.

FIG. 14.—SECONDARY SET OF LINES IN GRAIN. PRIMARY SET IN N.E. AND S.W. CORNERS WHERE THE SECONDARY LINES ARE ABSENT. THIS STRUCTURE ALSO OCCURS IN N-IZED IRON. FIRST TIME NOTICED IN ELECTRIC WELD. × 1200.

FIG. 15.—THE LARGE PLATE FOR SOME REASON HAS A WIDE GROOVE ON ONE SIDE. SLIP BANDS CROSS PLATE AND EXTEND ON OTHER SIDE. × 1200.

FIG. 16.

FIG. 17.

FIG. 16.—ENLARGED VIEW OF FIG. 15 SHOWING SLIP BANDS PASSING TO BOTTOM OF GROOVE, THROUGH PLATE, WHICH IS BROKEN IN NUMEROUS PLACES, RISING ON THE OTHER SIDE, AND CONTINUING THROUGH FERRITE. STEPS, DUE TO SLIPPING, ARE CLEARLY VISIBLE, THOUGH THEY ARE ROUNDED AND NOT LIKE SAW TEETH. × 2400.

FIG. 17.—SLIP BANDS PAY NO ATTENTION TO LINES, BUT SOMETIMES CROSS THEM, AND SOMETIMES STOP AT THEM, IRREGARDLESS OF THEIR DIRECTION. × 430.

necessary and that it varies greatly even under apparently the same conditions.

The etching has been done with a 2 per cent. solution of nitric acid in absolute alcohol and has been carried far enough to show the grain boundaries. The result of this heavy etching is to roughen the grains and produce structures that must be interpreted with care. However, as the object was to locate the grain boundaries, no attention has been paid to the structures shown by heavy etching unless they were checked by normal etching with both picric and nitric acids. In some cases, espe-

SPECIMEN 106.

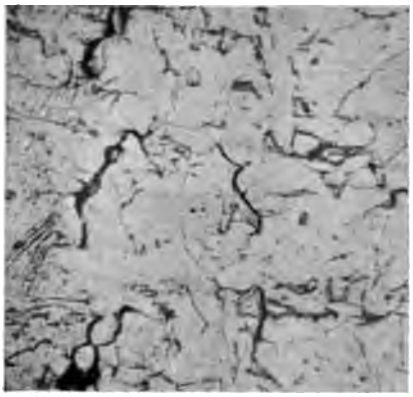

FIG. 18.

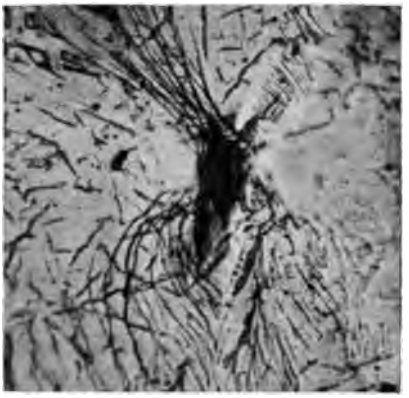

FIG. 19.

FIG. 18.—GRAIN BOUNDARY RUPTURES. × 430.
FIG. 19.—SLIP BANDS INTEGRATING AT DEFECT, PROBABLY A GAS POCKET. × 430.

In electric welds some of the grain boundary ruptures may be caused by invisible films of oxide. It is usually such ruptures that occur where there is no prior evidence of such films; but there are sometimes spots that seem porous or unsound, and oxide films may exist in their vicinity. It is well to distinguish between obvious oxide defects and grain boundary ruptures due to other causes. The first evidences of strain in any weld are distortions at visible defects.

cially where the grains are columnar, the orientation appears to be practically the same in many grains, and it is difficult to show the grain boundaries, even by very heavy etching. In such cases, heat tinting is sometimes of advantage, although this does not always show the difference between the grains.

While speaking of the grain boundaries, it may be well to call attention to a condition that is brought out by heavy etching, and which was particularly examined in specimen 4C4. Fig. 22 shows, at low power, the structure of the greater part of the weld; it seems quite clear that there are ferrite boundaries between the grains. The bodies of these grains appear to consist of ferrite with pearlitic lines along the cleavage planes; these are shown at higher power in Fig. 23. After deeper etching,

SPECIMEN 4C4.

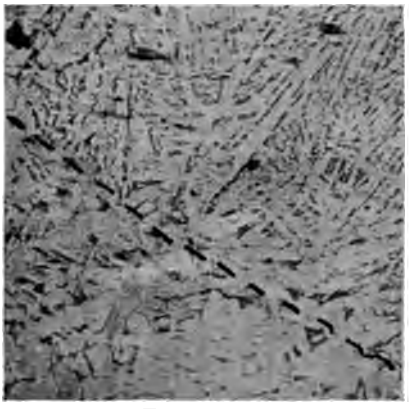

FIG. 20.

FIG. 21.

FIG. 20.—ON LINE OF WELD WHICH IS ABOVE DOTTED LINE. MUCH CARBON IS IN WELD. × 100.

FIG. 21.—ORIGINAL MATERIAL MELTED DOWN; IT IS NEARLY FREE FROM CARBON. VIEW IN BOTTOM OF WELD. × 100.

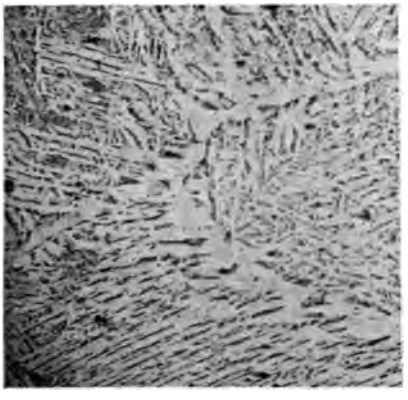

FIG. 22.

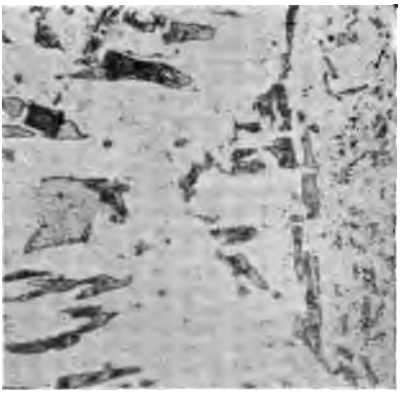

FIG. 23.

FIG. 22.—VIEW IN WELD AT TOP. MUCH CARBON IS PRESENT. FERRITE LINES ARE PROBABLY MAIN GRAIN BOUNDARIES. × 100.

FIG. 23.—ENLARGEMENT OF FIG. 22, SHOWING PEARLITE. × 430.

SPECIMEN 4C4.

FIG. 24.

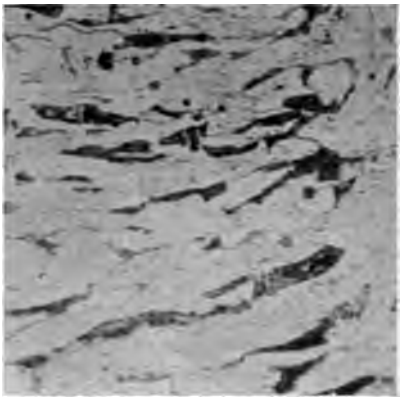

FIG. 25

FIG. 24.—HEAVIER ETCHING SHOWS GRAIN BOUNDARIES CLEARLY, WHERE ORIENTATION CHANGES, AND SECONDARY GRAIN BOUNDARIES IN S. E. CORNER. × 430.

FIG. 25.—SECONDARY GRAIN BOUNDARIES CLEAR; THE ORIENTATION IN THESE GRAINS IS PROBABLY THE SAME. × 430.

The author thinks these secondary grain boundaries are caused by the leaking out of solution of Fe_3C., time not being given for it to segregate. In some cases there is enough of it to make pearlite, in others only enough to appear as cementite, and in other cases the films may be ultramicroscopic.

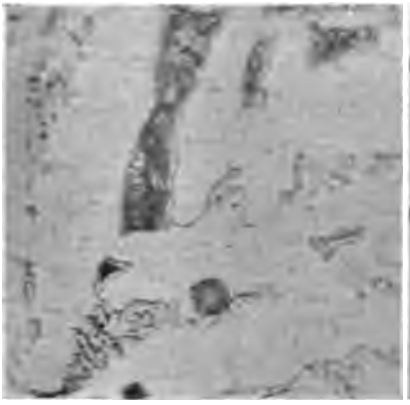

FIG. 26.

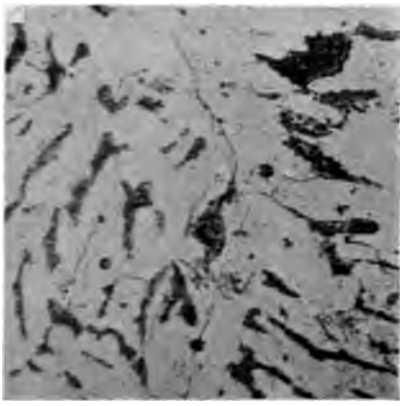

FIG. 27.

FIG. 26.—WHAT APPEARS AS PEARLITE AT LOW POWER HAS TWO ASPECTS AT HIGHER POWER. ONE VERY FINE GRAINED AND THE OTHER WITH DIVORCED CEMENTITE. × 1200.

FIG. 27.—UNDER STRAIN RUPTURE BEGINS AT MAIN GRAIN BOUNDARIES; SLIP BANDS IN ADJACENT FERRITE. × 430.

SPECIMEN 4C4.

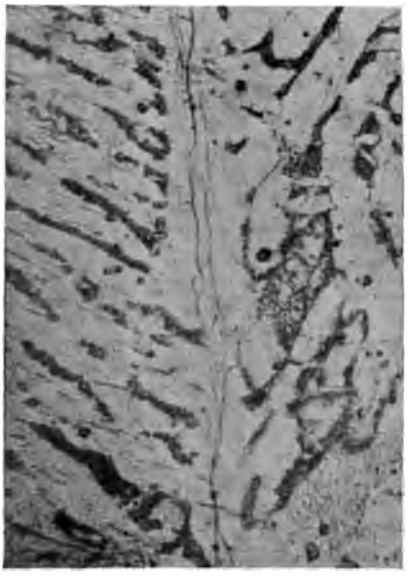

FIG. 28.

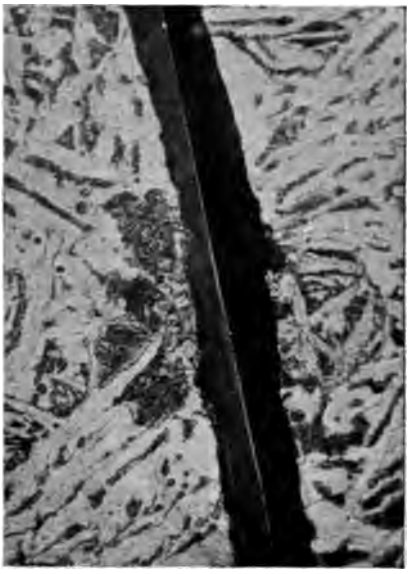

FIG. 29.

FIG. 28.—MAIN GRAIN BOUNDARY RUPTURE WITH ADJACENT SLIP BANDS. THIS IS VERY COMMON IN THIS WELD. × 430.

FIG. 29.—ABOUT ONE-THIRD OF A RUPTURE. BOTH SIDES PHOTOGRAPHED AS THEY WERE NOT IN THE SAME PLANE. RUPTURE PASSES AROUND PEARLITE AND IS ENTIRELY IN FERRITE. × 430.

FIG. 30.

FIG. 31.

FIG. 30.—GRAIN BOUNDARY RUPTURE IN BOTTOM OF WELD. × 430.

FIG. 31.—VERY FINE GRAIN BOUNDARY RUPTURE IN TOP OF WELD. × 430.

there appear boundaries of what are apparently smaller grains, and it is noticeable that these boundaries are clear and distinct where the orientations of the grains change, as shown in Fig. 24. In places where the orientation may be the same, as shown by the parallelism of the pearlite plates, it is usually difficult to see the grain boundaries, but sometimes this can be done, as in Fig. 25. The structure of what appears to be pearlite is shown at higher power in Fig. 26. There appear to be two varieties; in one the eutectoid is very fine, and in the other the cementite is clearly visible as such.

While this structure is from an oxyacetylene weld, electric welds are quite similar in appearance, though the eutectoid is probably not pearlite;

SPECIMEN 4C4.

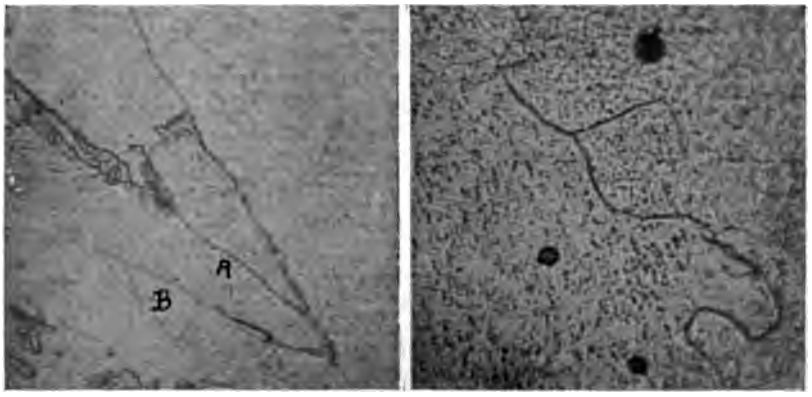

FIG. 32. FIG. 33.

FIG. 32.—FILMS OF CEMENTITE. THE ONE AT A IS VERY THIN. THE ONE AT B TAPERS TO NOTHING AND IS 0.00001 IN. THICK NEAR THE END. × 1200.

FIG. 33.—VERY FINE GRAIN BOUNDARY RUPTURE. MINUTE FILMS VISIBLE IN UPPER PART ARE APPARENTLY CEMENTITE. × 1200.

Further strain increased ruptures at ferrite grain boundaries both in size and number. Also the ruptures like Figs. 30, 31, and 33 increased. There were no ruptures through the pearlitic zone or along secondary grain boundaries.

Figs. 8 and 80. It should be remembered that oxyacetylene welds made with low-carbon material do not show this columnar structure nearly so much, and that these secondary grain boundaries, if they may be so called, have not been noticed, the grain size and structure being as in Figs. 46, 47, and 48. It should also be understood that almost any structure can be found in any weld and that the statements just made are the author's observation of the usual condition, to which there are exceptions.

The author feels that, in the present case, it is difficult to describe what occurs during the testing, and, therefore, has given most of the information with the photographs. His conclusions from his examina-

SPECIMEN Q

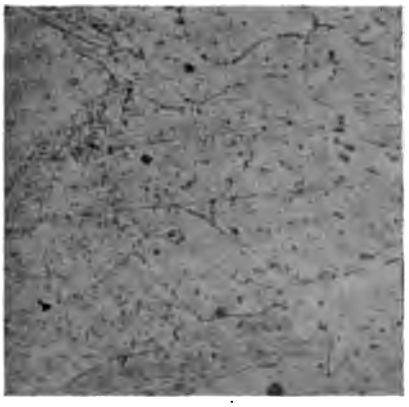

FIG. 34.

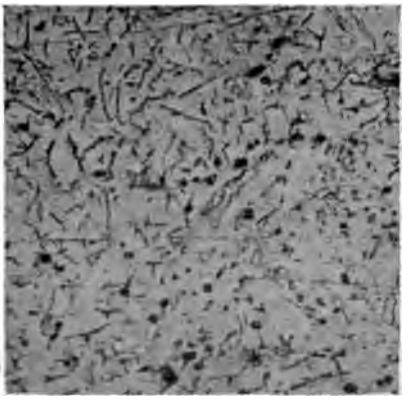

FIG. 35.

FIG. 34.—STRUCTURE IN LOWER PART OF WELD, WHERE GRAIN SIZE HAS BEEN ALTERED BY HEAT FROM UPPER LAYER. NO LINES IN GRAINS, BUT SOME SPOTS AT GRAIN BOUNDARIES. × 430.

FIG. 35.—USUAL STRUCTURE OF REST OF WELD. MORE OXIDE SPOTS THAN IN FIG. 34. × 430.

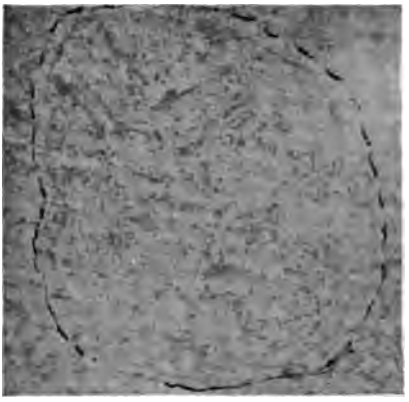

FIG. 36.

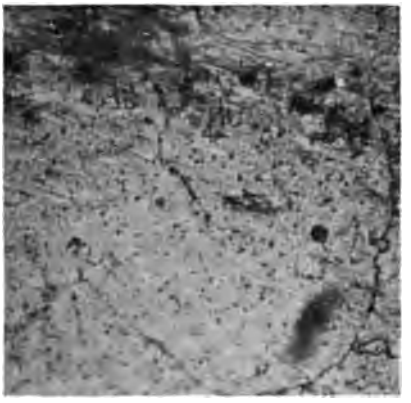

FIG. 37.

FIG. 36.—ONE OF NUMEROUS SIMILAR DEFECTS IN THIS WELD. VERY THIN FILM OF OXIDE AROUND GLOBULE OF METAL. LINE OF OXIDE JUST INSIDE DOTTED LINE. × 430.

FIG. 37.—SIMILAR DEFECT, WITH GAS POCKET OR PORES AT N.W. CORNER. × 430.

SPECIMEN Q.

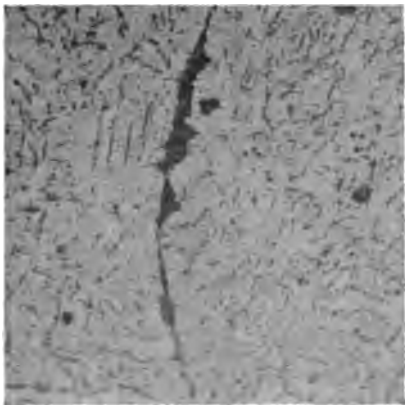

FIG. 38.

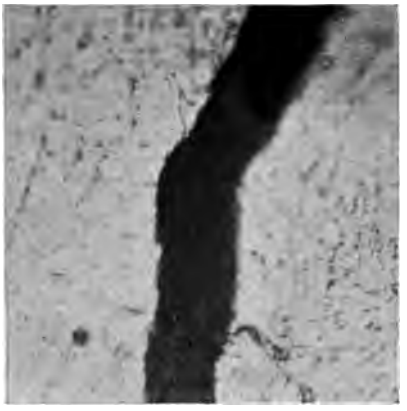

FIG. 39.

FIG. 38.—ANOTHER OXIDE FILM. ABOUT ONE-TENTH OF IT IS SHOWN. × 430.
FIG. 39.—FIRST STRAIN OPENS UP DEFECTS. ABOUT ONE-QUARTER OF DEFECT SHOWN. × 430.

FIG. 40.

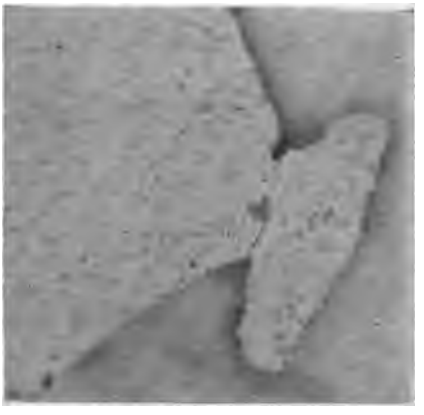

FIG. 41.

FIG. 40.—IN BOTTOM OF WELD, METAL BREAKING AWAY FROM SLAG. × 430.
FIG. 41.—RUPTURE AT LINE OF OXIDE SO MUCH BELOW METAL SHOWN THAT IT IS ENTIRELY OUT OF FOCUS. × 430.

SPECIMEN Q

FIG. 42.

FIG. 43.

FIG. 42.—SURROUNDING METAL IN FIG. 41 FOCUSED. OPENING OF CRACK AND ITS EXTENSION S.E. SHOWN. × 430.

FIG. 43.—FURTHER STRAIN OPENS DEFECTS MORE. SAME FIELD AT FIG. 39. × 430.

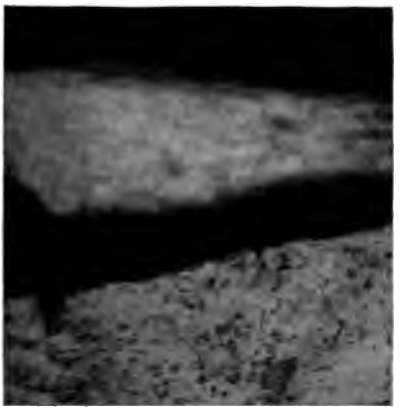

FIG. 44.

FIG. 45.

FIG. 44.—RUPTURES AT RIGHT ANGLES TO LINE OF STRAIN ARE SOMETIMES FOUND, ESPECIALLY AT DEFECTS. AT BOTTOM OF FIGURE IS METAL, THEN RUPTURE, THEN MORE METAL, OUT OF FOCUS, ITS TOP EDGE BEING THE TOP OF THE WELD. × 430.

FIG. 45.—SLIP BANDS AT UPPER END OF RUPTURE IN FIG. 39. × 430.

In this weld 16 defects were visible before applying strain, the area examined was ¼ by ⅜ in. It is perfectly possible to do much better than this, if proper electrodes and current are used, and care is taken.

This weld has been used as an illustration of what it is necessary to avoid; that is, freedom from even microscopic defects is necessary for the best work.

SPECIMEN RA1.

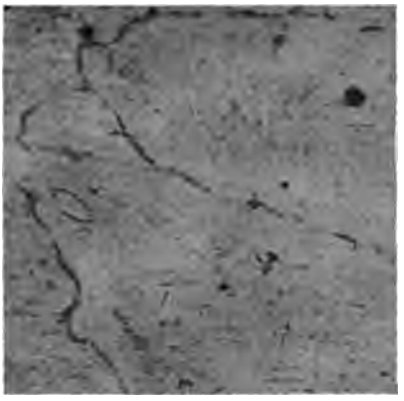

FIG. 46.

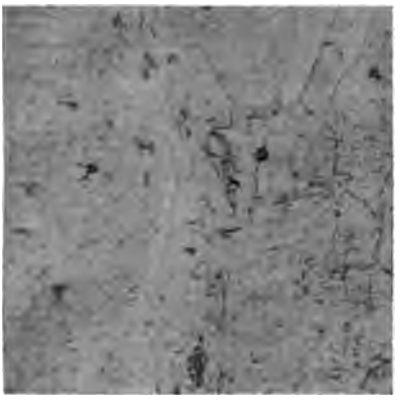

FIG 47.

FIG. 46.—GRAIN SIZE AND SHAPE. LARGER THAN IN CF2 AND NOT COLUMNAR. × 430.

FIG. 47.—STRAIN CAUSES SLIP BANDS, NOT GRAIN BOUNDARY RUPTURES. × 430.

FIG. 48.

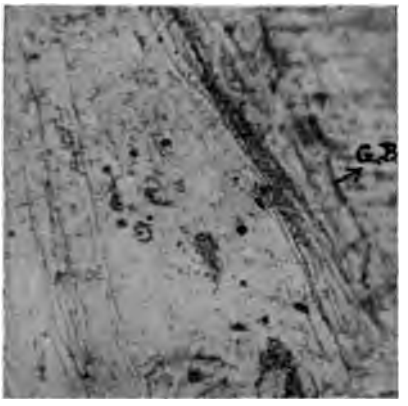

FIG. 49.

FIG. 48.—FURTHER STRAIN INCREASES NUMBER OF SLIP BANDS WHICH HERE CROSS GRAIN BOUNDARY. × 430.

FIG. 49.—SLIP BANDS INTEGRATE INTO GROOVES OR DIFFERENCES IN LEVEL, BUT MISS GRAIN BOUNDARIES AS SHOWN. × 430.

SPECIMEN RA1.

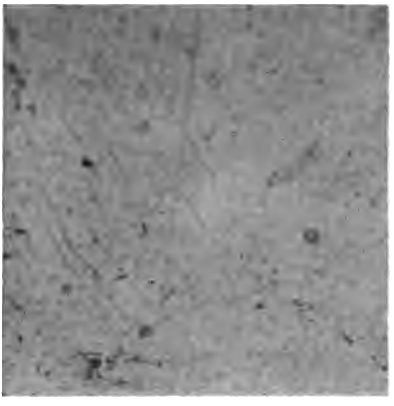

FIG. 50.

FIG. 51.

FIG. 50.—SLIP BANDS DO NOT HERE CROSS GRAIN BOUNDARY. THE NEARLY VERTICAL LINE IS A SCRATCH. × 430.

FIG. 51.—FURTHER STRAIN INTEGRATES SLIP BANDS INTO A GRAIN BOUNDARY RUPTURE. THIS IS THE ONLY CASE NOTICED WHERE RUPTURE WAS AT GRAIN BOUNDARY. × 430.

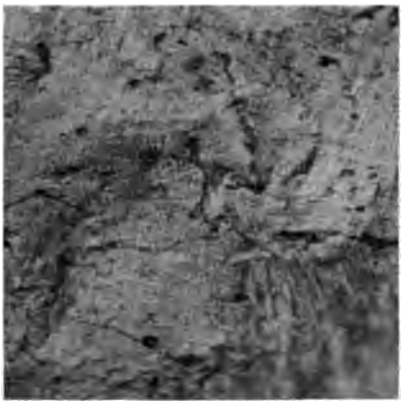

FIG. 52.—WITH FURTHER STRAIN SURFACE BECOMES ROUGH AND UNEVEN DUE TO MANY SLIP BANDS, ETC. THERE IS NO EVIDENCE OF GRAIN BOUNDARY RUPTURES. × 430.

SPECIMEN RE2.

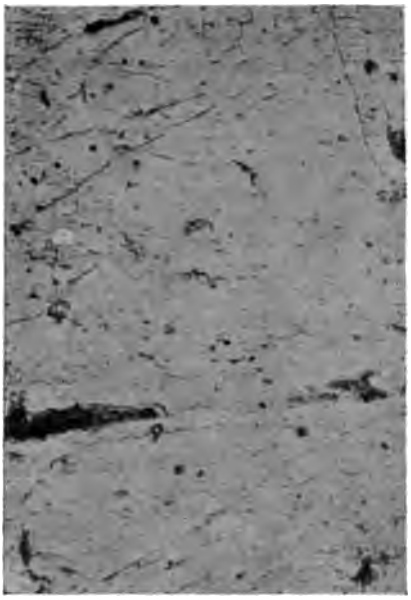

FIG. 53.

FIG. 54.

FIG. 53.—GRAIN SIZE AND SHAPE SIMILAR TO CF2. QUITE A LITTLE PEARLITE. GRAINS COLUMNAR. × 430.

FIG. 54.—GRAIN BOUNDARY RUPTURES APPEAR WITH SLIGHT STRAIN. × 430.

FIG. 55.

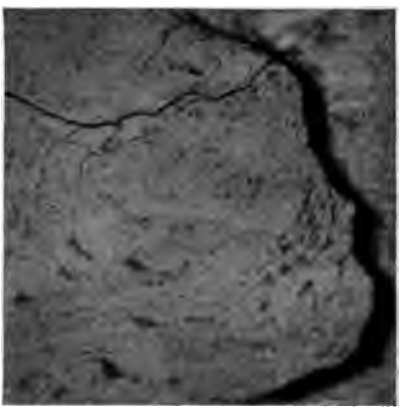

FIG. 56.

FIG. 55.—SLIP BANDS APPEAR ALSO, BUT ARE VERY FINE. × 430.

FIG. 56.—LARGE DEFECT APPEARED AT FIRST STRAIN. FURTHER STRAIN CAUSED IT TO OPEN MORE, AND STARTED SMALLER CRACK SHOWN. BOTH AT GRAIN BOUNDARIES. × 430.

tions, not only of the specimens mentioned but of a number of others, are as follows:

1. The first evidence of strain in any weld is at whatever visible defects may exist in the weld, such as films of oxide around the grains, or around small particles of metal as in electric welds. This is well shown in Figs. 7, 13, 39, and 40. The latter defects are very common in electric welds, as in Figs. 36, 37, and 38; and the films are sometimes very thin, as in Fig. 36. It will be noticed from the table that the welding materials low in carbon appear to give a much less columnar structure in gas welds than the others, compare Figs. 46, 5, 53, and 60; this, however, is difficult to prove by photographs. Also, in electric welds, the structure is usually

SPECIMEN RE2.

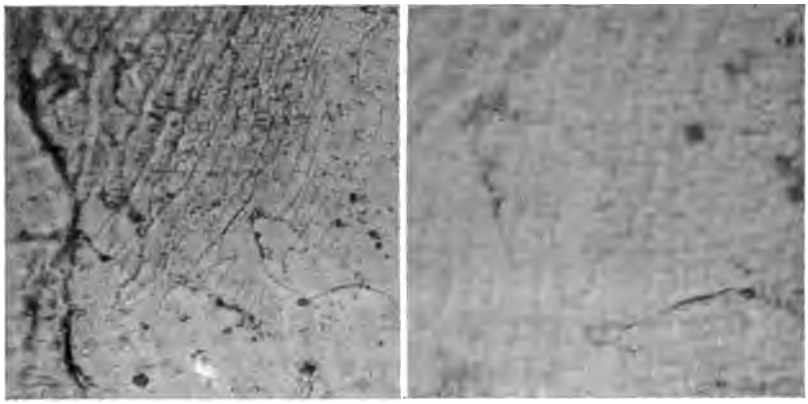

FIG. 57. FIG. 58.

FIG. 57.—SLIP BANDS INCREASE RAPIDLY WITH FURTHER STRAIN AND SOMETIMES CROSS GRAIN BOUNDARIES. × 430.

FIG. 58.—ENLARGED VIEW OF FIG. 57. NINE SLIP BANDS CROSS GRAIN BOUNDARY. ALL COULD NOT BE PHOTOGRAPHED AT ONCE. × 1200.

RE2 seemed more brittle than CF2; that is, more ruptures appeared, and less strain was needed to cause them. Slip bands seemed to increase more rapidly than in CF2, but less rapidly than in RA1. As in CF2, grain boundary ruptures cause most of the distortion.

more columnar than in gas welds; although variations can be noticed here also. It would appear that the rapid cooling of electric welds is responsible for this and that the slower cooling of gas welds probably makes the grains more nearly equiaxed. Again, any material that preserves the pearlitic structure in a gas weld, such as vanadium or manganese, seems to produce more columnar grains. Aluminum, in considerable quantities in electric welds, makes the grains excessively columnar, the structure being shown in Fig. 80, and very large, their columnar appearance in the fracture being noticeable to the naked eye.

2. If there are no defects in the weld, the first appearance of distortion varies with the kind of weld and the material with which it is made.

SPECIMEN CF2.

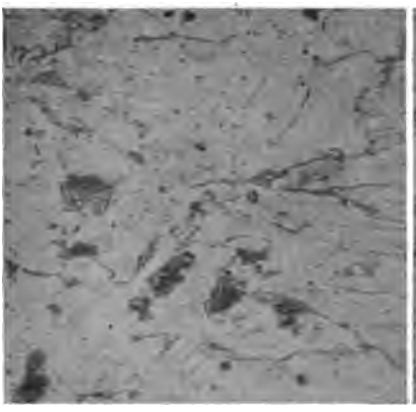

FIG. 59.

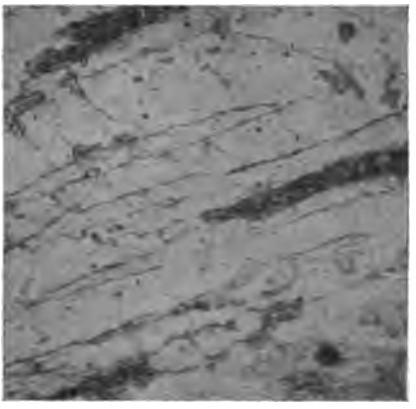

FIG. 60.

FIG. 59.—GRAIN SHAPE AND SIZE VARIES FROM THIS TO FIG. 60. × 430.
FIG. 60.—COLUMNAR GRAINS OF FERRITE AND PEARLITE. × 430.

In both of these there is considerable pearlite and cementite along grain boundaries, due to vanadium in welding rod. The grains are also small compared with RA1.

FIG. 61.

FIG. 62.

FIG. 61.—GRAIN BOUNDARY RUPTURE AFTER MODERATE STRAIN. × 430.
FIG. 62.—GRAIN BOUNDARY RUPTURE WITH SLIP BANDS AT END OF CRACK RUNNING PARALLEL TO GRAIN BOUNDARY. SEE FIGS. 64, 66 AND 72. × 430.

SPECIMEN CF2.

FIG. 63.

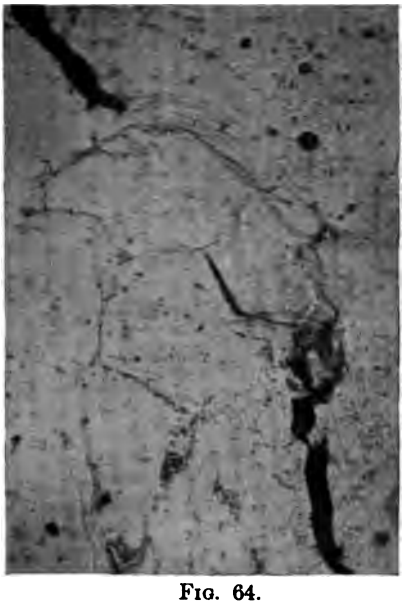

FIG. 64.

FIG. 63.—ANOTHER GRAIN BOUNDARY RUPTURE. × 430.

FIG. 64.—FURTHER STRAIN ENLARGES RUPTURES AND SHORT STRAIGHT RUPTURE BEGINS TO CROSS GRAIN BOUNDARY. × 430.

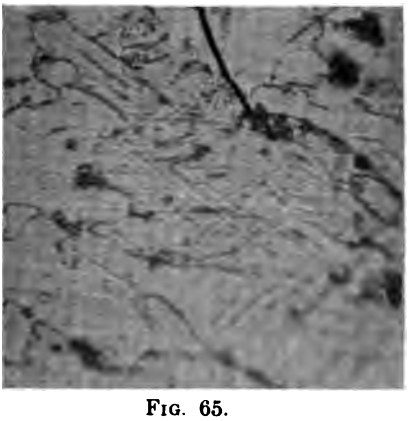

FIG. 65.

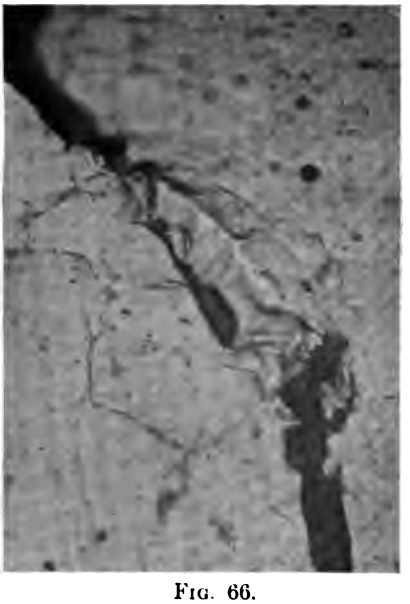

FIG. 66.

FIG. 65.—ALSO SMALLER RUPTURES OCCUR, SLIP BANDS INCREASE. TWO SETS SHOWN IN SAME GRAIN AT CENTER OF FIELD WHICH IS NEAR BOTTOM OF WELD. × 430.

FIG. 66.—WITH FURTHER STRAIN THE RUPTURES BECOME WIDER AND IN THIS CASE HAS CROSSED THE GRAIN IN CENTER. SUCH A CROSSING OF THE GRAIN IS RARE. *SLIP BANDS ALSO INCREASE.* × 430.

SPECIMEN CF2.

FIG. 67.

FIG. 68.

FIG. 67.—GRAIN BOUNDARY RUPTURE IN FIG. 63 EXTENDS AND SLIP BANDS APPEAR IN GRAIN. × 430.

FIG. 68.—FURTHER RUPTURES OCCUR AT GRAIN BOUNDARIES. × 430.

FIG. 69. × 430.

FIG. 70. × 1200.

Grain boundary rupture extending into pearlite. In larger view, end of pearlite at rupture shown to be below surface and as slip bands do not appear to be large enough to equal the width of the crack, it would seem that slip has also occurred in the pearlite.

In oxyacetylene welds, made with pure iron or low-carbon material, in which practically all the carbon is burnt out during the welding, and where there is but little if any manganese or vanadium, slip bands appear in the grains first, as in the various photographs of specimen RA1. These slip bands at times cross the grain boundaries, as in Fig 48, and at other times stop at the grain boundaries, as in Fig. 50. They appear to increase in number and form cracks, as in Figs. 49 and 51; but there is seldom any evidence of rupture at the grain boundaries; in

SPECIMEN CF2.

FIG. 71.

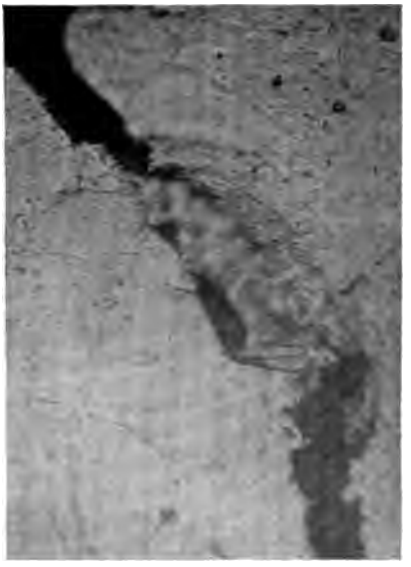

FIG. 72.

FIG. 71.—GRAIN BOUNDARY RUPTURE PASSING THROUGH PEARLITE, THE ONLY CASE NOTICED. × 1200.

FIG. 72.—FURTHER INCREASE IN WIDTH OF RUPTURE AND IN SLIP BANDS; SEE FIGS. 62, 64, AND 66. × 430.

While many slip bands appear in CF2, they are very fine and distortion increases at grain boundaries rather than by slip bands.

Fig. 49 the grain boundary is shown to the right of the set of slip bands. Fig. 51 shows what at first appears to be a grain boundary crack, but much of it is really nothing but an accumulation of very fine slip bands. Eventually the surface becomes a mass of slip bands and folds, as in Fig. 52; after which it becomes impossible to determine where the final rupture occurs though it is doubtless along some of the slip bands.

3. In all electric welds and in oxyacetylene welds made with material containing considerable carbon, say 0.4 per cent., or in which other elements such as manganese and vanadium are in sufficient quantity to maintain a pearlitic structure even with lower carbon, which structure

SPECIMEN EWN—NORMALIZED FROM 900° C.

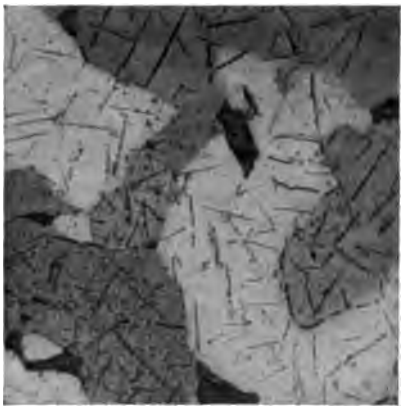

FIG 73.

FIG. 74.

FIG. 73.—GENERAL STRUCTURE. HEAT-TINTED AFTER ETCHING. DARK PATCHES AND LINES ARE PURPLE. LINES CLEARER THAN IN WELD. COLUMNAR STRUCTURE DESTROYED. LINES APPEAR SOMETIMES AT GRAIN BOUNDARIES. × 430.

FIG. 74.—ONE OF THE LARGE PURPLE PATCHES WITH DARKER PURPLE MARKINGS INSIDE RESEMBLING MARTENSITE IN SHAPE. THESE MAY CONTAIN SOME CARBON AND APPEAR IN NEARLY ALL PATCHES. LINES SHOW CLEARLY ON GRAIN BOUNDARIES. × 1200.

SPECIMEN EWQ—HEATED FOR 1 HR. AT 900° C. AND QUENCHED IN COLD WATER.

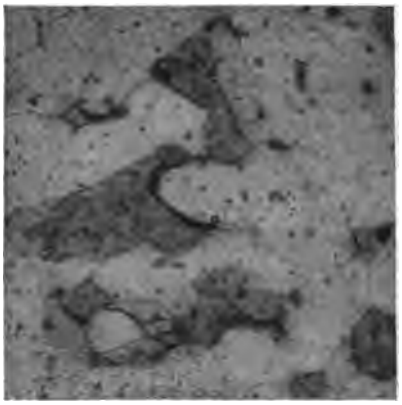

FIG. 75.

FIG. 76.

FIG. 75.—GRAINS OF PEARLITIC APPEARANCE ARE POSSIBLY AN IRON-IRON NITRIDE EUTECTOID. LINES IN GRAINS HAVE DISAPPEARED. × 430.

FIG. 76.—EUTECTOID APPEARS VERY FINE GRAINED. SAME AS FIG. 75, BUT × 1200.

SPECIMEN CBE1.

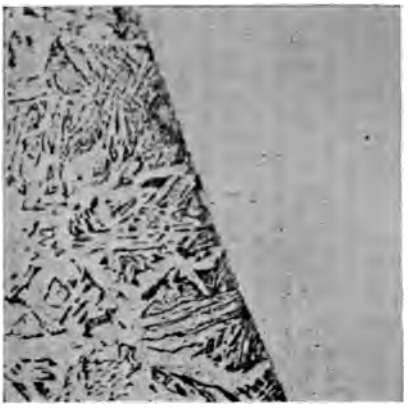

FIG. 77.

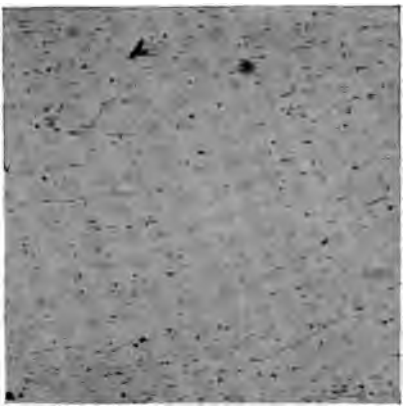

FIG. 78.

FIG. 77.—MATERIAL ON RIGHT, ORIGINAL MELTED DOWN AND DECARBURIZED. MATERIAL ON LEFT, ORIGINAL ALTERED IN STRUCTURE BY THE HEAT. × 100.

FIG. 78.—FILMS AROUND GRAINS IN ORIGINAL MATERIAL MELTED DOWN. × 100.

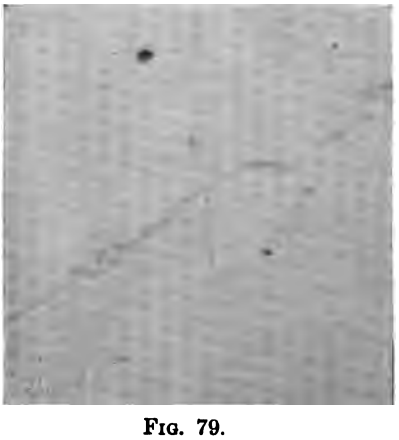

FIG. 79.

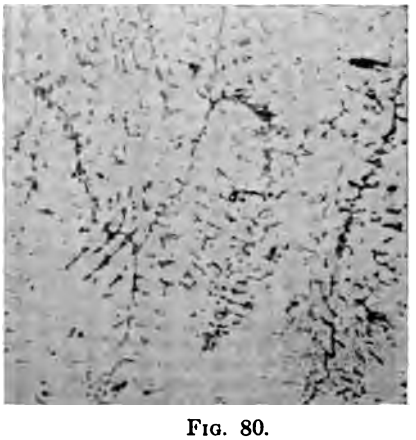

FIG. 80.

FIG. 79.—SAME AS FIG. 78, BUT × 1200. NOTE ONE SPOT WHERE FILM DOES NOT SHOW. IT IS PROBABLY TOO FINE TO BE SEEN. THE FILMS ARE PROBABLY Fe_3C.

FIG. 80.—LARGE COLUMNAR GRAINS IN ADDED MATERIAL. THIS IS USUAL STRUCTURE OF THIS WELD. × 100.

is also quite columnar and resembles that of electric welds, the path of rupture is invariably along the grain boundaries.

SPECIMEN BS1.

FIG. 81.—STRUCTURE OF LIGHT SPOTS IN COLUMNAR GRAINS. THEIR NATURE IS NOT KNOWN BUT POSSIBLY THEY ARE Fe_3C PRESERVED BY THE ALUMINUM. × 1200

These conclusions apply only to welds as made and not to those that have been heat treated. Also, the path of rupture due to shock or

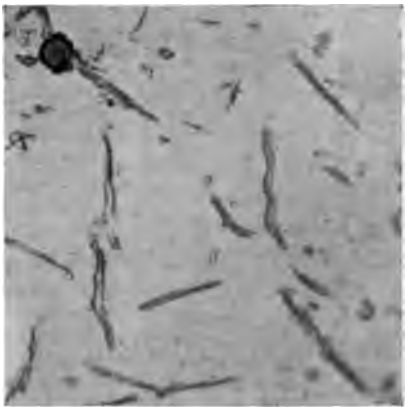

FIG. 82.—LINES IN TOP OF OXYACETYLENE WELD RESEMBLING THOSE IN ELECTRIC WELDS, BUT LARGER AND THICKER BECAUSE OF SLOWER COOLING. THEIR NATURE IS NOT KNOWN. THEY MAY BE CAUSED BY ABSORPTION OF NITROGEN AT THE VERY HIGH TEMPERATURE BY THE MELTED METAL. × 1200.

alternating stress may possibly be different from that due to tensile stress. The author will make some tests along these lines later.

These facts are very interesting and evidently have a bearing on the brittleness of certain welds. For instance, it is well known that electric welds are comparatively brittle, but the author has never seen the statement that oxyacetylene welds made with material containing 0.15 per cent. carbon and 0.15 per cent. vanadium, or that containing 0.15 per cent. carbon and 0.50 per cent. manganese are just as brittle as electric welds, if not more so. Such oxyacetylene welds are also much higher in tensile strength than welds made with low-carbon material free from manganese and vanadium. Also, welds made with 0.4 per cent. carbon steel with normal amounts of the other usual elements are brittle.

All brittle welds resemble each other in that the structure is more or less columnar and that they contain impurities such as pearlite, cementite, and nitride of iron(?). According to the usual understanding, during the solidification or cooling from high temperature of a metal, impurities are rejected to the grain boundaries. Humfrey, in a very interesting paper[2] on the intercrystalline fracture of steel has shown that intercrystalline rupture can be caused by overheating, and that proper heat treatment restores the tendency to rupture across the grains in the normal way. The author is of the opinion, from rough tests he has made, that the brittleness in gas welds at least can be removed by similar heat treatment. He is not so sure of electric welds, because their heat treatment, at least in any ordinary way, does not remove the line structure, as shown in Fig. 73, in which, in many cases, the lines lie either on or very close to the grain boundaries. Fig. 74 shows some of these lines directly on the grain boundaries.

There are also, in welds so treated, and which have been heat-tinted after etching, patches lying at the grain boundaries; these are purple when the surface of the specimen is a light brown. The lines also are purple. These purple patches usually, but not always, contain zigzag lines of a still darker purple, as shown in Fig. 74. The author does not know the nature of these zigzag lines and purple patches, but inasmuch as similar patches in unannealed welds darken when boiled in sodium picrate, he thinks that they contain carbon. The purple patches are probably nitride of iron, which heat-tints purple when pure.

The quenching of electric welds appears to produce an entirely different structure, as shown in Figs. 75 and 76. The grains here, as in the the normalized specimen, have lost their columnar nature, as would be expected, and the lines have entirely disappeared, the nitride of iron apparently forming a eutectoid with some of the iron. These photographs show the necessity of further investigation. Andrew has shown that nitrogen, in sufficient quantity, suppresses the critical points,

[2] *Carnegie Schol. Mem.*, Iron and Steel Inst. (1912) **4**, 80–107.

so that it is a question as to the exact conditions under which these changes occur.

It seems quite clear that the only way to account for the brittleness of welds is to assume that it is caused by films of material at the grain boundaries, the nature of these films differing in different welds. In specimen CBE1, made by the carbon-arc process, there is one spot where the original material was melted down and decarburized by the heat, in which there are very faint films around the grains, as shown in Figs. 78 and 79. These films are evidently cementite trapped at the grain boundaries by the rapid cooling. They are 0.00002 in. thick in places, and in spots where they cannot be seen it would seem entirely reasonable to suppose that they exist, but are ultra-microscopic. Similarly, specimen 4C4 shows in many places very thin films of cementite of the same or even less thickness as shown in Figs. 32 and 33.

It would not seem probable that in metallic electrode welds such films would be of cementite, as the carbon is almost entirely burned out; they may be nitride of iron. It is also possible that, as Humfrey suggests in the paper referred to, they are oxide of iron. Fig. 6 shows comparatively thick films of iron oxide in an oxyacetylene weld, in which there were nests of these films in many places. Evidently crystallization cannot proceed through these films, whatever their nature, nor does it seem possible that amorphous material could exist there. It is quite probable that where the films are cementite, they could be absorbed by heat treatment. This could occur when the films are nitride of iron and probably is impossible when they are oxide of iron. It seems quite plausible that these films could be of ultra-microscopic thickness; and as all of them are brittle, they would be very weak under shock and probably under alternating stress, although their thinness might account for the high tensile strength of the welds in which they exist, on the same principle that a thin film of glue is stronger than a thick film. It might be mentioned that metal electrode welds have sometimes a tensile strength as great as 70,000 lb. per sq. in. (49.19 kg. per sq. mm.) and would probably have this always if they were sound. Oxyacetylene welds made with low-carbon material have much greater ductility and resistance to shock because of the absence of these films; and it appears probable that, within limits, the purer the material the more ductile such a weld will be.

While the author questions whether anything but circumstantial evidence can ever be adduced for his belief in the presence of these films, the facts are that, under welding conditions, very thin films that are visible do exist; and that with welds containing impurities, the breakage is always at the grain boundaries where it is known that impurities collect. So he sees no reason to doubt their existence and believes that they are responsible for the brittleness of such welds.

DISCUSSION

A. M. CANDY,* Pittsburgh, Pa.—I think more stress should be laid upon the question of welding with the carbon electrode, which we ordinarily call graphite arc welding. Mr. Miller's photograph indicated that the original parent metal had been decarburized to a marked degree. It would be interesting to know the carbon content of the parent metal, and whether or not Mr. Miller has made any investigations to determine if the metal is always decarburized or whether in some instances we actually have carbon injected into the weld, which has been the understanding that has been quite prevalent.

There were one or two other points that Mr. Miller brought out concerning which I desire to ask one or two questions: The illustrations that he presented were obtained with what, I suppose, would be termed a real low-carbon steel welding wire, which gives in the deposited metal grains that are more or less symmetrical, whereas the Roebling electrode, which is a mild steel of higher carbon content, apparently gives a columnar structure. I do not know which of these structures is more desirable, but assuming that the structure produced by the really low-carbon steel wire is the more desirable, would it not be true that the very low carbon content would, probably owing to high temperatures that prevail in the arc, result in a greater amount of oxidation and probably result in either carrying into the weld more oxide or producing more oxide on the surface as the operator progresses?

The short arc is an advantage in all cases. Whether Mr. Miller's trouble is oxide or nitride, the short arc helps him out, because the closer the end of the electrode is to the work the less chance there is for either oxygen or nitrogen to get in. And whichever is the cause of the trouble, the short arc keeps it out by lessening the time in which the steel is molten and the amount of air that gets in. It also makes sure that the molten metal will not fall into the crater. Everything is in favor of the short arc.

How can that short arc be obtained? With an ordinary direct-current machine there is nothing to limit a man to a short arc; but every good operator will limit himself to a short arc. There is one direct-current system that necessitates holding a short arc due to the low voltage; other systems provide relays and other moving parts to snuff out the arc when it reaches a certain voltage, but in the case of the machine that develops low voltage, penetration is lost, for that depends not only on the amount of current but also on the line voltage. The low-voltage machine (35 volts) is not so satisfactory as a higher voltage machine (60 volts) because the operator will experience more difficulty in holding the arc As a result, the arc will be broken more frequently, materially reducing

*General Engineer, Westinghouse Elec. & Mfg. Co.

the operator's production and producing a larger percentage of slag inclusion due to the oxidation always occurring on the surface of the crater where the arc breaks. Because of the larger number of arc interruptions, also, the metal in the crater is always noticeably porous. On the whole, therefore, instructing an operator as to the advantages of holding a short arc and giving him a reasonable open-circuit voltage (60 volts) will produce the most satisfactory results because the effect of momentary long arcs cannot be detected in the welding work.

GEORGE F. COMSTOCK,* Niagara Falls, N. Y. (written discussion†). This paper by Mr. Miller is a very interesting study of fusion welds, and his conclusion that films at the grain boundaries cause brittleness in such welds certainly seems well founded by his work, and gives the paper practical value as well as interest. There are a few minor details of the paper that might be criticized, and a few suggestions also that I would like to make. In the first place, nitric acid in ethyl alcohol has been found in our laboratory to work much better in developing grain boundaries than the same acid in absolute alcohol, and it does not roughen the ferrite grains so much.

The author's belief that all patches that darken when boiled in sodium picrate must contain carbon is erroneous, as there is no doubt that nitride is also darkened by this treatment; and it is hard to see why cementite films should exist in decarburized metal full of oxide spots, as seems to be assumed by the author in some cases. The limitations of the ordinary processes of polishing these metal specimens do not seem to be appreciated by the author, for when one stops to consider that the final polished surface is obtained by eliminating scratches by surface flow of the metal, and that in the preliminary grinding it is rather easy to erode brittle non-metallic inclusions, there should be no surprise if occasionally a very fine crack or brittle film is covered up in polishing so that it is invisible in spots.

I would like to ask Mr. Miller, as a matter of theoretical interest, for his opinion as to why the needles appear only at the surfaces of oxy-acetylene welds, and not in the interior as is the case with electric welds.

I would also like to ask if it would not seem reasonable to suppose that the intercrystalline brittleness and excessively columnar structure noticed in welds containing manganese, vanadium, or aluminum may be due to oxides of these elements which oxidize before iron, depositing in a finely divided state along the grain boundaries? In this connection it seems unfortunate that titanium has not been tried as a means of collecting and removing these finely divided oxides, in view of the general experience with this element in the ordinary processes of steel manufacture.

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† Received Feb. 24, 1919.

ZAY JEFFRIES,* Cleveland, Ohio (written discussion†).—Mr. Miller's work on the path of rupture in fusion welds is fundamental. He is to be congratulated on the thoroughness with which he outlined and executed the experimental work. Following are some observations based on the data included in the papers and elsewhere.

The iron nitride needles greatly increase the elastic limit of the ferrite-nitride complex. This is proved by the fact that the elastic limit of the weld material is very much higher than the elastic limit of pure iron. The effect of the nitride needles in increasing the elastic limit seems to act in a manner somewhat similar to the increase in elastic limit of ferrite by the presence of cementite.

If the welded-in material containing nitride needles is markedly more deformable under pressure than under tension, an increase in strength at the grain boundaries, at which place rupture first occurs, besides increasing the strength, would make the welds more ductile in tension. The most ductile crystalline grains would act as if brittle if they were held together by a cement having a breaking load less than the elastic limit of the grains.

In the old austenite grain boundaries, which show themselves as wide bands of nearly pure ferrite as indicated in Mr. Miller's Fig. 28, there may be two reasons for fracture through the ferrite. First, and this is suggested by Fig. 28, the boundary between the old austenite grains may contain some impurities, which remain on cooling through the critical range. These impurities would cause weakness and would facilitate rupture at the old austenite boundary. On the other hand, the old austenite grain boundary might actually have been sound, that is, free from objectionable inclusions and the elastic limit of the interior of the grain might have been so increased by the addition of the nitride and small quantities of carbon that the weakest point in tension would have been the pure ferrite network.

It seems from the photo-micrographs that the presence of oxide at the grain boundaries, or at least at the boundaries of these included spherical masses, is certain. On the other hand, as indicated in Fig. 73, it seems certain that iron nitride also collects at the grain boundaries. It is not certain, however, that this small amount of nitride weakens the grain boundary to an extent at all comparable with the decrease in strength caused by films of oxide. Iron nitride seems to possess metallic characteristics, whereas the oxide possesses non-metallic characteristics. The first might act as a strengthener, as is suggested by the fact that in no single case did Mr. Miller succeed in producing fracture along a nitride needle in the interior of a grain, but it is well known that a layer of iron oxide in the interior of a grain or at the boundary between two grains is a plane of weakness.

* Director of Research, Aluminum Castings Co. † Received Mar. 24, 1919.

Microstructure of Iron Deposited by Electric Arc Welding

BY GEORGE F. COMSTOCK,* A. B., MET. E., NIAGARA FALLS, N. Y.

(New York Meeting, February, 1919)

THESE notes should be considered as a further discussion of Mr. S. W. Miller's paper on "Some Structures in Steel Fusion Welds."[1] In that paper and the resulting discussion, several conflicting opinions were expressed as to the identity of the needles or small crystals present in oxyacetylene and electric welds. An interesting specimen of an electric weld examined by the writer seems to afford new evidence regarding the needles, that may be worth presenting at this time. The specimen consisted of a piece of $\frac{3}{8}$-in. (9.5-mm.) steel plate, about $2\frac{1}{4}$ by $1\frac{3}{4}$ in. (5.7 by 4.5 cm.) in size, on which some iron had been deposited by arc-welding to a depth of about $\frac{3}{16}$ in. (4.7 mm.). The weld appeared to be a very good one, with no noticeable break or boundary between the plate and the deposited iron.

Sections for microscopic examination were cut in three planes at right angles to each other through the deposited metal, two of these sections also extended through the steel base. After polishing in the usual way, the microscope showed that the deposited metal was full of very small, round, gray, oxide spots, while the steel plate contained quite a number of alumina inclusions and a little slag. The sections were etched with nitric acid in ordinary alcohol, and structures similar to those described by Mr. Miller were then displayed. Fig. 1 shows the typical appearance of the deposited metal, which was quite uniform in all three of the planes examined. The round dark-gray spots are oxides, and the pale angular crystals are evidently what Mr. Miller called cementite and Mr. Jeffries martensite, while Mr. Ruder and Professor Boylston think that they might be nitride. The boundary between this structure and the steel base was sharply defined under the microscope. Fig. 2 shows the structure of the plate just under the weld; this is almost martensitic and resembles a coarse-grained casting. This coarse cast structure soon became finer as the distance from the weld increased, finally becoming very fine, with numerous small particles of sorbite. This very fine structure, shown in Fig. 3, became coarser in turn and finally merged into the original normal structure of the steel plate, shown

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[1] *Trans.* (1918) **58**, 700.

in Fig. 4. Mr. Miller mentioned the coarse structure occurring next to welds, but not the refinement of the original grain farther away, although the latter seems equally characteristic, at least when soft steel is welded.

With a view toward acquiring further information regarding the identity of the pale crystals in the deposited metal, small sections of the sample were cut and one was annealed at 500° C., or below the critical point of the steel, for 2 hr., and cooled in lime. Another was annealed at 900° C., or above the critical point, for 4 hr., and cooled slowly in the furnace. The first annealing did not affect the different zones in the steel plate appreciably but caused the deposited metal, after etching as before with nitric acid, to appear full of darkened needles in place of the pale angular crystals. These needles, shown in Fig. 5, looked somewhat like sorbite deposited along cleavage planes, giving support to the view that the pale crystals might be martensite. The longer annealing above the critical point produced the structure shown in Fig. 6 after the same nitric acid etching. Some needles are still seen here, but most of them seem to have coalesced into pale irregular masses with darkened centers, looking much like the segregated cementite found, as Mr. Miller noted, in annealed low-carbon steel wires and sheets. The coarse and fine zones in the steel plate disappeared after this hotter annealing, and both ferrite and pearlite were coarsened, as shown in Fig. 7. At the boundary between the deposited metal and the steel base, a narrow zone existed where the needles had invaded the ferrite in the steel without merging with the pearlite that was present. The pearlite stopped abruptly at the upper boundary of this zone, which fact, together with the presence of alumina inclusions in the mixed zone, showed that it was the needles and not the pearlite that had migrated by diffusion. Fig. 8 shows the structure of this narrow zone, with the needles and pearlite present together yet sharply distinct from each other.

This was an unexpected development from the annealing experiments, for if the pale crystals and the needles were cementite, martensite, or some of its decomposition products, they would be expected to merge with the pearlite in the drastically annealed specimen shown in Fig. 8; and if they were nitride, they should disappear according to the German article quoted in my previous discussion.[2] Mr. Ruder, however, said that the nitride was hard to remove from steel, so it is probable that the German work was erroneous on this point, and the nitride theory as to the identity of the needles thus appears most plausible. As a further check, the remainder of the original sample submitted was filed so as to remove all the deposited metal with as little contamination as possible from the steel base. These filings, representing metal shown in Fig. 1, were analyzed for carbon by combustion and only 0.04 per cent. of this

[2] *Trans.* (1918) **58**, 720.

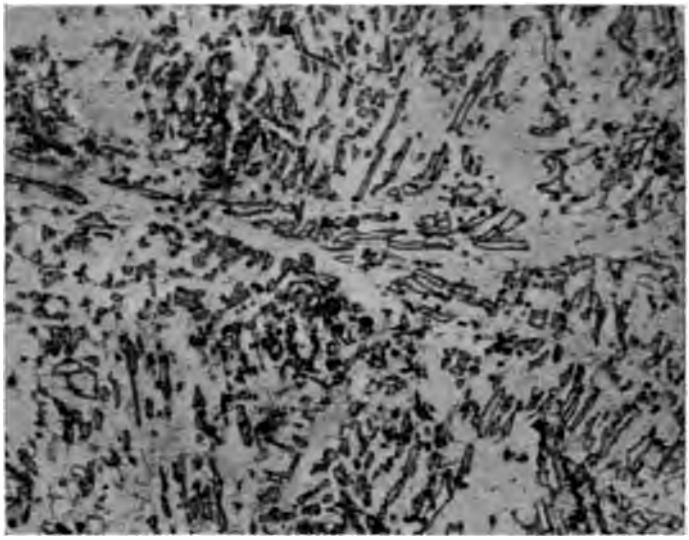

FIG. 1.—DEPOSITED METAL OF WELD BEFORE ANNEALING, ETCHED WITH NITRIC ACID. × 400.

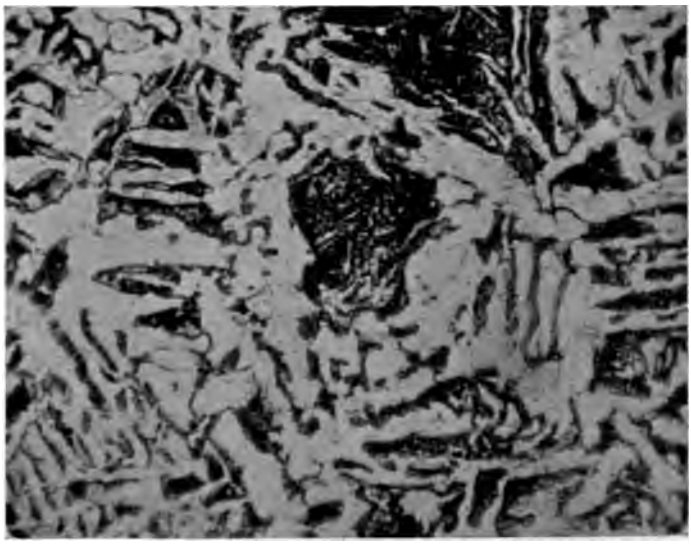

FIG. 2.—STEEL PLATE JUST BELOW WELD, ETCHED WITH NITRIC ACID. × 400.

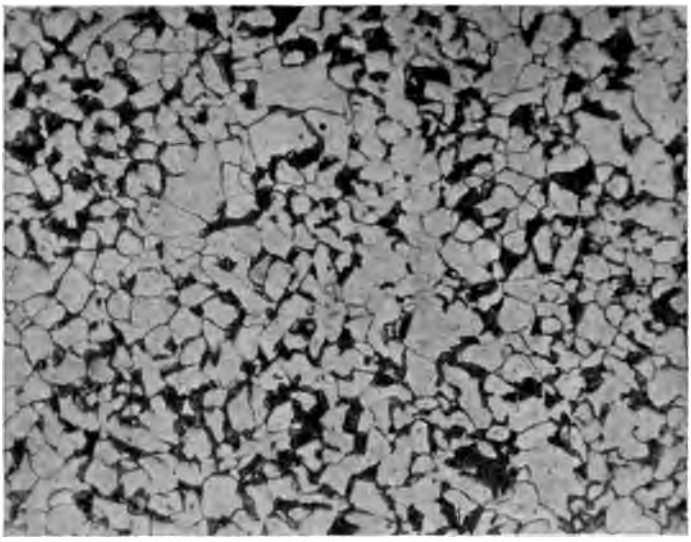

FIG. 3.—STEEL PLATE ABOUT 3/16 IN. BELOW WELD, ETCHED WITH NITRIC ACID. × 400.

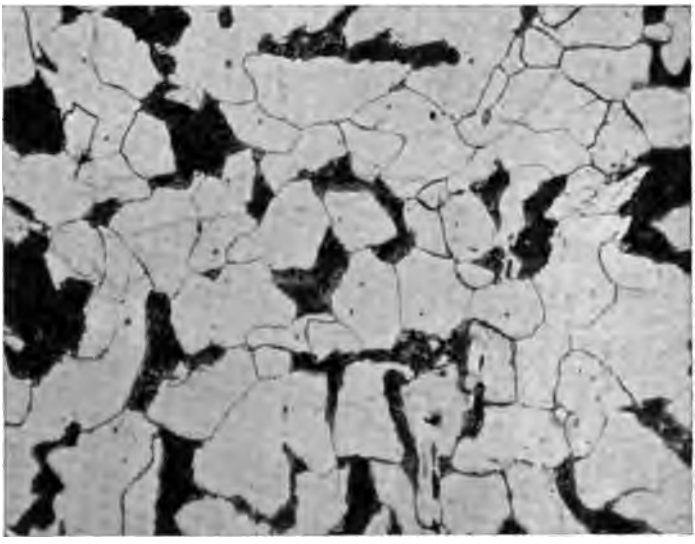

FIG. 4.—STEEL PLATE BEYOND INFLUENCE OF WELD, ETCHED WITH NITRIC ACID. × 400.

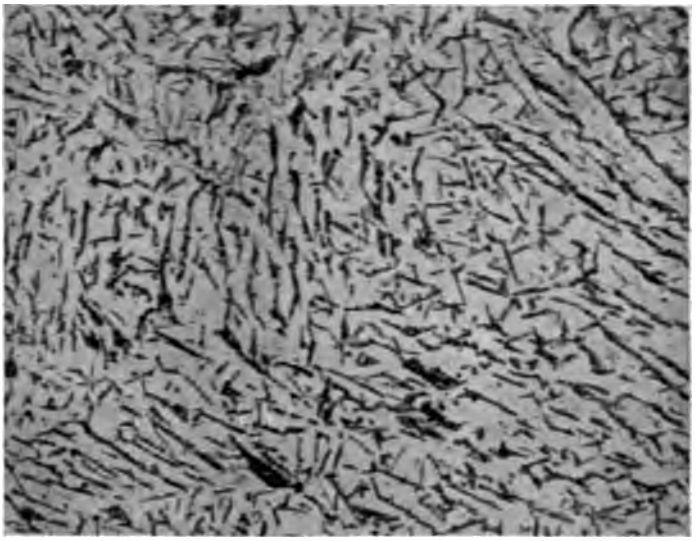

FIG. 5.—DEPOSITED METAL OF WELD AFTER ANNEALING AT 500° C. FOR 2 HR., ETCHED WITH NITRIC ACID. × 400.

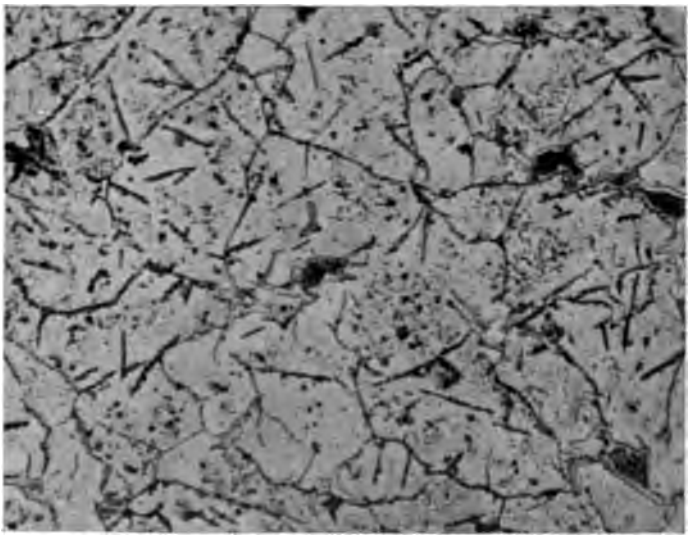

FIG. 6.—DEPOSITED METAL OF WELD AFTER ANNEALING AT 900° C. FOR 4 HR., ETCHED WITH NITRIC ACID. × 400.

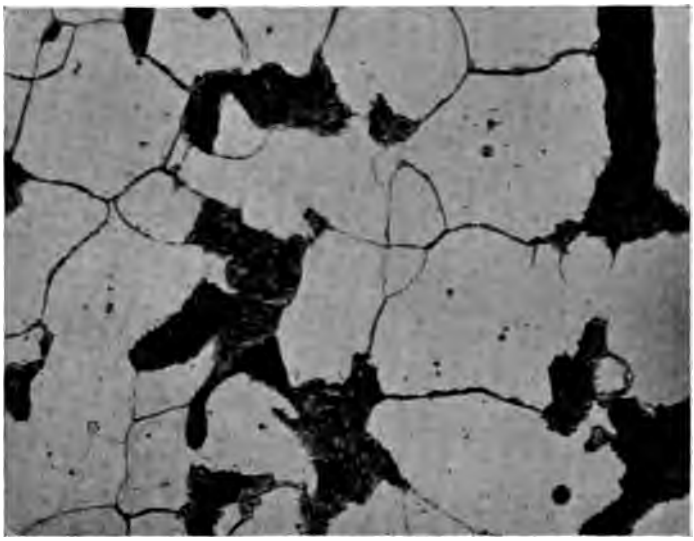

FIG. 7.—STEEL PLATE BELOW ABOUT 1/64 IN. FROM WELD AFTER ANNEALING AT 900° C. FOR 4 HR., ETCHED WITH NITRIC ACID. × 400.

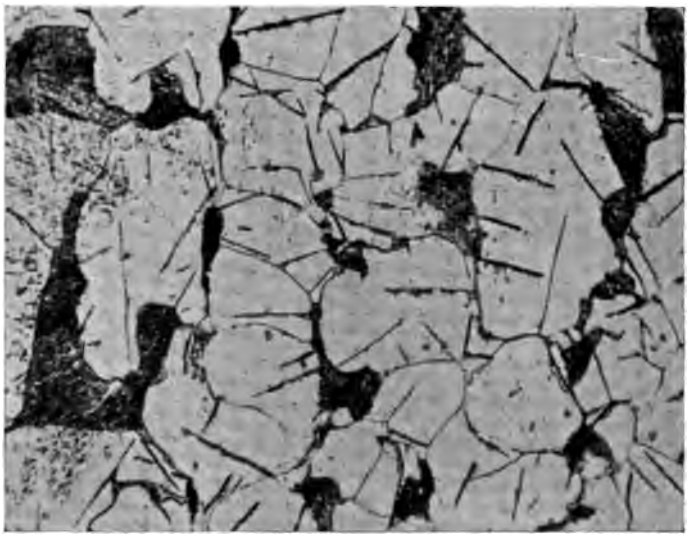

FIG. 8.—STEEL PLATE JUST BELOW WELD AFTER ANNEALING AT 900° C. FOR 4 HR., ETCHED WITH NITRIC ACID. × 400.

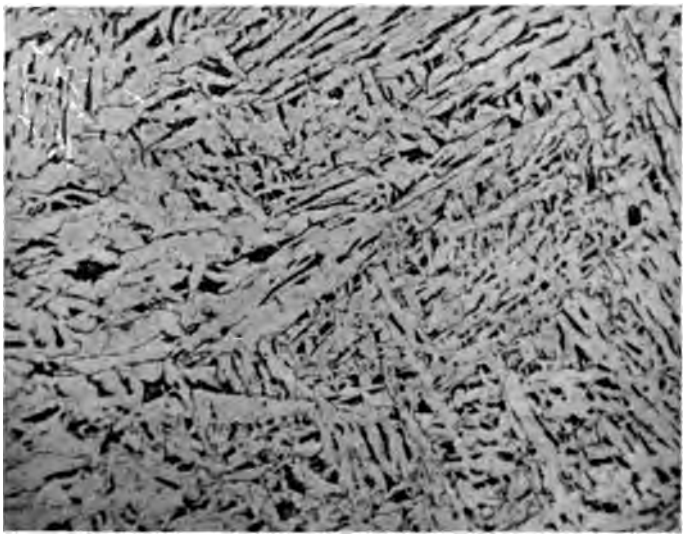

FIG. 9.—DEPOSITED METAL OF WELD BEFORE ANNEALING, BOILED 45 MIN. IN ALKALINE SODIUM PICRATE. × 400.

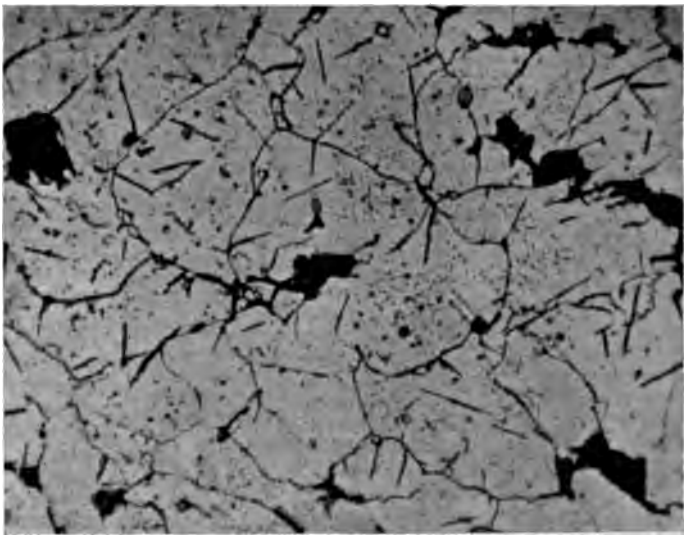

FIG. 10.—DEPOSITED METAL OF WELD AFTER ANNEALING AT 900° C. FOR 4 HR., BOILED 10 MIN. IN ALKALINE SODIUM PICRATE; SAME SPOT SHOWN IN FIG. 6. × 400.

element was found. Certainly, therefore, the pale crystals cannot be cementite, and judging from the amount of decomposition product shown in Fig. 5, more carbon would have to be present if this were all assumed to have come from martensite. It is thus most probable that the pale crystals in Fig. 1, the needles in Figs. 5, 6, and 8, and the pale irregular areas with dark centers in Fig. 6 are all various forms of nitride of iron. Since these structures are all typical of those that Mr. Miller showed, the writer believes that, in most cases at any rate, the structures that he found were due to nitride of iron which assumed various forms according to the speed of cooling after welding. The disappearance of the needles in his annealing experiment agrees with the change shown between Figs. 5 and 6, which were cut from the same specimen containing only 0.04 per cent. carbon, but after different heat treatments. Of course, these nitride crystals may not always be pure when found in welds and may very likely contain some carbide in solution, but evidently there are cases, such as the one here described, where the crystals or needles are chiefly nitride rather than cementite.

Etching with boiling alkaline sodium picrate was tried on a specimen from the original sample, with the result that the pale crystals were not changed after the usual 10-min. treatment. After boiling for 45 min., however, during which time the solution became more concentrated and the boiling point therefore higher, the crystals were darkened, as shown in Fig. 9. The specimen which had been annealed at 900° C. needed only the usual 10-min. etching to darken its nitride particles, with the result shown in Fig. 10, which should be compared with Fig. 6. This shows that this method of etching darkens not only cementite and sulfide inclusions, but also nitride crystals if continued long enough.

Summarizing, the results of this investigation show that the pale crystals or needles typical of steel fusion welds may occur very abundantly in iron containing only 0.04 per cent. carbon, and that when annealed thoroughly and slowly cooled the needles will diffuse into steel from the welded metal but do not merge with the pearlite of the steel. The conclusion, therefore, is that the pale crystals or needles are not cementite or martensite or any similar carbide product, but most probably nitride of iron. Mr. Ruder's statement that nitride is not easily removed by heating is corroborated, though the chilling effect of the welding on the steel can be removed easily by annealing. From the above conclusion and the sodium picrate etching tests, it appears further that boiling in this solution will not differentiate between nitride and carbide or sulfide in a polished specimen of steel.

The writer wishes to express his thanks to Mr. E. Wanamaker, electrical engineer of the Rock Island Lines, for the sample investigated, and to Mr. P. Munnoch chemist of the Titanium Alloy Manufacturing Co., for the chemical analysis.

DISCUSSION

E. E. THUM,* Salt Lake City, Utah (written discussion†).—While the present paper is confined exclusively to the microstructure of fusion welds, it may be of interest to draw attention to certain appearances existing in resistance welds such as spot or butt-welded joints produced by the Thomson process. For instance, a butt weld between two pieces of carbon tool steel (carbon about 0.90 per cent.) will show concentric ellipsoids of the various metarals from pearlite to martensite. The surfaces of demarcation are quite well defined, the center of the ellipsoid being at the intersection of the axis of the piece and the plane of the weld. Often these unannealed welds show a troostitic center; evidently this core was not quenched as rapidly as the surrounding shell. Such specimens are beautiful samples for instruction purposes.

An annealed butt weld of this kind will show a very well defined central ellipsoid of decarburized materials. It is interesting to consider the reason why this change takes place. It must be understood that the heat is on the metal for a very short time, possibly 30 sec. It is therefore evident that the decarburization, or migration, of the carbon takes place with extreme rapidity. Since no blowholes are evident in a well-made butt weld, the oxidizing air and the products of combustion must migrate with the highest speed; second, the carbon might migrate back into the stock; or, last, a large proportion of the pearlite may be extruded with the hot metal, which forms a fin at the plane of weld.

Examining these possibilities in detail, it is not probable that such deep-seated decarburization, extending with maximum effect to the very center of a 2-in. round bar, is done by reaction with the oxygen of the air. This would postulate entrance of oxygen and egress of carbon oxides in large quantity without leaving behind traces in the form of oxides or blowholes. On the other hand, these blowholes are quite numerous in spot welds formed in such a manner that plastic metal is not extruded in any quantity. Backward migration of carbon into the stock was investigated both chemically and microscopically without finding any intermediate zone enriched in carbon. The final possibility, namely, that the cementite is plastic at a lower temperature than its associated ferrite and is preferentially extruded, was not investigated because facilities were lacking to make a butt weld in a neutral atmosphere so that the extruded metal should not be largely burned. This is an attractive possibility but is hard to reconcile with the observed fact that fine-grained pearlite goes into austenitic solution almost instantly at temperatures above the recalescent point.

The microscopic appearance of unannealed spot welds in low-carbon

* Western Editor, *Chem. & Met. Eng.*

† Received Feb. 28, 1919.

tank steel and other mild-steel pieces is quite different. The ellipsoidal kernel is in this case sorbitic. Immediately surrounding this is a relatively thin shell of eutectiform appearance, gradually merging into the unaffected stock. A number of blowholes can easily be distinguished by the eye near the center of the weld. These are evidently due to a reaction between the rusty scale and the iron carbide existing in the steel. Microscopic investigation shows these holes to be surrounded by decarburized metal and often associated with iridescent areas. These blowholes may be found at distances as far back as 0.125 in. (3 mm.) from what was the original surface of contact. Since the electric current was on for 10 sec. or even less, it is seen that migration of various substances is rapid at this point as well. Of course, it is possible that local fusion may have taken place in this region.

FIG. 11.—EUTECTIFORM STRUCTURE IN SPOT-WELDED MILD-STEEL PLATES.

The eutectiform structure immediately overlying the sorbitic center was puzzling in the extreme. My assistant who was polishing the specimen first reported that he had developed the most beautiful pearlite he had ever seen. Evidently the area could not be pearlite, however, since the stock contained less than 25 points of carbon, and there were no evidences of migration. More than one competent observer has identified it as martensite but on being informed as to the composition of the metal has admitted that he would not expect such well-developed and large areas in low-carbon steels. Also, martensitic structure should not occur between well-defined zones of sorbite and pearlite.

After considerable study and investigation, the following conditions seemed to apply to the formation of this peculiar structure. It occurred in spot welds only; from the nature of this operation it was in metal that had been subjected to a considerable compressive stress at the time of welding. The material had evidently been heated to and probably above

the transformation range, and had also been subjected to a drastic quenching, due to the rapid absorption of the localized welding heat by the surrounding cold metal. Further, one must assume almost instantaneous absorption of the ferrite crystals in the neighboring pearlite with the formation of a uniform mass of austenite.

Imagine, then, the sudden and localized formation of a body of such austenite by electric heating, and the crystals of austenite so formed then subjected to a compressive stress sufficient to develop slip bands along its octahedral cleavage. Electric current is now suddenly interrupted, the hot spot at the weld is drastically quenched in the surrounding cold metal, and the austenite precipitates its dissolved ferrite. At the center of the weld this decomposition produces well defined sorbite. Surrounding the sorbitic volume, however, is a shell of fractured austenite which has been heated but slightly above its transformation range or is cooled at such a speed that the agglomeration of the precipitated ferrite is not complete. But bear in mind that each austenite crystal has been broken up into a multitude of slabs and the excess constituent contained in these slab-like masses is expelled to the boundaries of the slab and has no time to coalesce into the larger masses of ferrite that characterize slowly cooled low-carbon steels. On polishing and etching, therefore, the nuclear planes of the austenite crystals are each marked by a thin layer of cementite crystals, the whole under the microscope giving rise to roughly parallel striations branching at about 60°. In some specimens, the striations are not very straight but the feature of parallelism is somewhat too prominent to allow them to be identified as pearlite. Other sections are given an appearance strongly resembling deeply etched austenite. Again, as already noted, the eutectiform structure taken by itself would probably be identified as martensite. Enough has been given, however, to explain my reason for identifying this appearance as extremely delicate Widmannstättian structure. Can it be possible that complete diffusion of carbide and iron can take place with such extreme speed?

W. E. RUDER,* Schenectady, N. Y. (written discussion†).—Although I am still unprepared to say that these lines are nitride, it is certain that they are due to the presence of nitrogen, so I prefer to call them N lines until their exact nature is certain. The mistake has frequently been made in concluding that, inasmuch as the specimen showed none of the N lines, nitrogen was not present. As I have already pointed out in a previous paper, and as Mr. Comstock has clearly brought out in his micrographs, the lines are only evident upon comparatively slow cooling. and then only when nitrogen is present in amounts approximating 0.08 per cent. or over.

* Metallurgist, General Electric Co. † Received Mar. 1, 1919.

FIG. 12.—TREATED ELECTROLYTIC IRON. NH_3. 700° FOR 24 HR.

FIG. 13.—1000° C. IN A CLOSED TUBE FOR 1 HR. QUENCHED IN WATER.

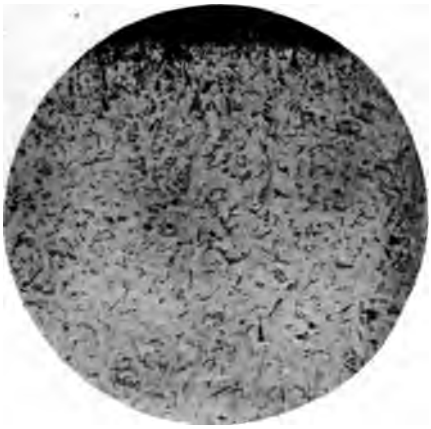

FIG. 14.—1000° C. IN A CLOSED TUBE FOR 1 HR. QUENCHED IN WATER. 600° C. FOR 1½ HR. COOLED SLOWLY.

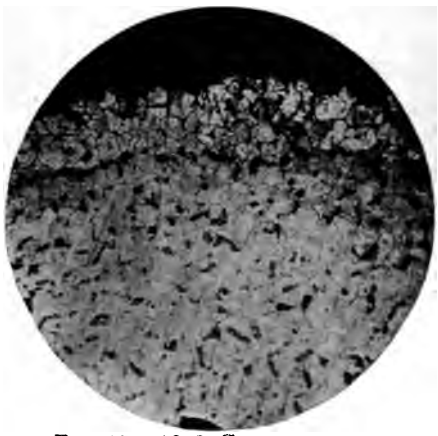

FIG. 15.—1090° C. IN A CLOSED TUBE FOR 1 HR. QUENCHED IN WATER. 600° C. FOR 1½ HR. COOLED SLOWLY. 1000° C. FOR 1 HR. COOLED SLOWLY.

FIG. 16.

I have been making a study of the appearance of nitrogenized pure iron and low-carbon steel under varying conditions of heat treatment. Unfortunately, this work is not quite complete but will be published at an early date. I am giving herewith some micrographs that will illustrate my present point. These are of pure electrolytic iron treated in ammonia at 700° C. for 24 hr. and cooled with the furnace. Fig. 12 represents the original structure after the nitrogen treatment. The dark area through the center is a groove cut to identify the spot examined. The similarity between Fig. 13 and unannealed weld section is apparent. I have found that nitrogen may occur in at least three different forms when associated with pure iron: as a solid solution in the iron, and hence unseen by the microscope; as widened grain boundaries; and as needles or N lines. Any or all of these forms may be present at the same time. One of the most satisfactory ways of studying these nitrogen-bearing samples is by heat tinting. The diffusion of the nitrogen into the ferrite of the plate stock noted by Mr. Comstock, and shown in Fig. 8, is quite evident in all well-annealed weld sections.

As for the disappearance of the nitride on heating, later experiments made to determine this point confirmed my original statement, referred to by Mr. Comstock, that the nitrogen is not so easily gotten rid of as the German papers would lead us to suppose. In my experiment, it took about 2 hr. heating at 1200° C. in a vacuum to clear the specimens of nitrogenized iron and weld of all evidence of nitrogen. This disappearance does go on at lower temperatures, but at a much slower rate. It is certain that an ordinary annealing will reduce the nitrogen present only a very slight amount.

Work on nitrogen-bearing low-carbon steels is complicated by the apparent combination of the nitrogen and the carbon. I do not believe that a cyanide is formed, but, at any rate, there seems to be some sort of combination. This gives a slightly different appearance to the specimen. Such steels always have structures resembling very closely those of unannealed welds. The lines are not so sharply defined but take rather the form of parallel bands, giving the grains a rectangular appearance as shown in Fig. 16.

ZAY JEFFRIES,* Cleveland, Ohio (written discussion†).—Inasmuch as I was responsible for the statement that the needles in electric welds might be a form of martensite, I am glad to take this occasion to announce my agreement with Messrs. Ruder and Comstock that these characteristic needles are due to the presence of nitrogen. Mr. Ruder's work in this field has been so thorough that the conclusion admits of little argument. On the other hand, the idea should not prevail that martensite does not occur in electric welds high in carbon because these often show well defined martensitic structures.

* Director of Research, Aluminum Castings Co. † Received Mar. 24, 1919.

GEORGE F. COMSTOCK (author's reply to discussion*).—Mr. Thum's description of the effects of butt welding and spot welding on the microstructure of steel are interesting and timely. The speed with which the changes noted by Mr. Thum must take place is indeed remarkable, but the writer has noted a similar very rapid diffusion of carbide in annealing at a rather high temperature certain samples of soft sheet steel containing massive cementite segregated on a small scale. The massive cementite in such steel can be dissolved and distributed very quickly if the proper temperature is attained. Mr. Thum's explanation of the structure he found in the spot welds seems entirely reasonable, but the mystery of the decarburized material in the butt weld is one that might well be the subject of further research. The possibility of decarburization by air might perhaps be worthy of a little more consideration than he gives it, for the oxides formed would be gaseous on account of the presence of carbon, and hence not visible in the microscope, and the effect of the pressure might be to squeeze the carbon oxides, which would normally form blowholes, out of the metal or possibly into such small volumes as not to be noticed. The removal of carbon by liquation and squeezing into the fin would be an effect similar to the "bleeding" of a steel ingot that is rolled while the center is still liquid, giving a soft low-carbon interior in the finished section while the segregate was removed in cropping the pipe. This effect in Mr. Thum's butt welds could, however, only be obtained if the weld was heated to near the melting point of the steel, and if the carbon was considerably segregated in the bars at the start, for the first effect of the high temperature would be to promote diffusion rather than the contrary segregation and liquation.

It is gratifying to learn from Mr. Ruder's remarks that his researches have resulted in no disagreement with the few points brought out by the writer's experiments. In addition to the three forms in which nitrogen may occur in iron that he mentions, it would seem that a fourth form is often found. This is apparently a eutectoid, like pearlite, and sometimes almost as coarse as the massive cementite seen in soft steel sheets that have been annealed for a long time. One of the things needed in the metallographic study of steel is an etching reagent that will distinguish definitely between nitride in this form and the more common pearlite or cementite.

* Received Apr. 28, 1919.

Welding Mild Steel*

BY H. M. HOBART,† NEW YORK, N. Y.

(New York Meeting, February, 1919)

THIS paper deals principally with investigations undertaken by the Welding Research Sub-committee of the Welding Committee of the Emergency Fleet Corporation. The general object of the investigations has been to extend the use of welding in the construction of merchant ships and, specifically, to provide a definite basis for obtaining the greatest economy and efficiency in employing welding in place of riveting in the construction of the hulls of such ships.

Composition of Ship-plate Steel

The chemical composition of the steel employed in such hull construction varies with the thickness of the plates. Through the courtesy of Mr. H. Jasper Cox, of Lloyd's Register of Shipping, the following information may be given concerning the kind of steel plate employed in American shipyards in 1918 for the hull construction of merchant ships. Lloyd's requirements do not relate to the chemical composition. They require a tensile strength of 58,000 lb. per sq. in. (40.75 kg. per sq. mm.) for their lower limit and 72,000 lb. per sq. in. (50.59 kg. per sq. mm.) for their upper limit. For the information of the Committee, Lloyd's obtained from their surveyors at various works data of the carbon content, which is as follows:

Works		Carbon Content for Plates ¾ In. Thick	Carbon Content for Plates 1 In. Thick
A		0.14	0.23
B		0.14	0.25
C		0.19	0.25
D		0.20	0.30
E	Upper limit	0.30	0.35
	Lower limit	0 24	0.29
F	Upper limit	0.25	0.30
	Lower limit	0.21	0.27
G	Upper limit	0.25	0.35
	Lower limit	0.22	0.28

*Report of research under the joint auspices of the National Research Council and the Emergency Fleet Corporation.

†Chairman, Welding Research Sub-committee.

Works H employs: Shapes about ½ in. thick, 0.24 per cent. to 0.30 per cent carbon; shapes about 1 in. thick, 0.28 per cent. to 0.35 per cent. carbon. Small shapes such as: 2½ × 2¼ × ¼-in. angles, about 0.15 per cent. carbon; 4 × 4 × ⅞-in. angles, about 0.20 per cent. carbon.

From several tons of ½-in. (12.7 mm.) plate from the yard of the Chester Shipbuilding Co., which was employed in making many sample welds in an investigation designated the Wirt-Jones Tests, seven analyses were made at the Bureau of Standards. The maximum and minimum percentages of each of the impurities for these seven samples were as follows:

	Maximum Per Cent.	Minimum Per Cent.
Carbon	0.25	0.24
Manganese	0.46	0.45
Phosphorus	0.043	0.039
Sulfur	0.031	0.027
Silicon	0.052	0.024

For this material the Bureau of Standards reports:

Yield point, 37,900 lb. per sq. in.; ultimate tensile strength, 63,600 lb. per sq. in.; elongation in 2 in., 38.6 per cent.

The following manufacturer's data apply to about 10 tons of ½-in ship plate supplied by the Worth Steel Co., of Claymont, Del., and to be used for testing electrodes:

Chemical Analysis (Ladle Analysis)

	Per Cent.
Carbon	0.29
Manganese	0.37
Phosphorus	0.015
Sulfur	0.032

Physical Properties

Tensile strength, lb. per sq. inch	67,400
Elongation, per cent. in 8 in	25.25

Another lot of about 1½ tons of ½-in. and 1-in. ship plate kindly furnished to the Committee by the American Steel & Wire Co. for the purposes of the Committee's researches was analyzed by the Electrical Testing Laboratories with the following result, four analyses being made for each thickness:

	Maximum Per Cent.		Minimum Per Cent.	
	½ In.	1 In.	½ In.	1 In.
Carbon	0.24	0.28	0.22	0.26
Manganese	0.44	0.53	0.40	0.47
Phosphorus	0.033	0.033	0.028	0.027

The specifications of the American Society for Testing Materials for structural steel for ships (serial designation A 12–16, p. 98, A. S. T. M. *Standards*, 1918) are in abstract as follows:

Phosphorus (acid steel), not over 0.06 per cent.; phosphorus (basic steel), not over 0.04 per cent.; sulfur, not over 0.05 per cent.; tensile strength, between 58,000 and 68,000 lb. per sq. in.; elongation, minimum per cent. in 8 in. 1,500,000/tensile strength.

From these data we have a good idea of the kind of steel in connection with which it was the Committee's first and specific task to investigate welding.

Two kinds of welding are under investigation at present: fusion welding and spot welding. These are totally different kinds of welding. The fundamental difference is that while in fusion welding no pressure is employed, the success of spot welding is entirely dependent upon the application of both heat and pressure. For the spot welding of thick plates, the required pressure is very great.

Fusion Welding

The term fusion welding is employed to cover gas welding and electric-arc welding.

Gas welding is usually effected by simultaneously fusing with an oxyacetylene flame the material at and near the surfaces which it is desired to join, and some material (which is usually similar in composition) in the form of a rod, the tip of which is subjected to the heat of the flame. The oxyacetylene flame is directed with one hand and the welding rod is manipulated with the other hand.

Electric-arc welding may be subdivided into several classes. The two broadest classes are: Carbon-arc welding and the metallic-arc welding.

In carbon-arc welding, an arc is established between a carbon or graphite electrode (usually a graphite electrode) and the two pieces of steel it is desired to join. This graphite electrode is manipulated with one hand and a welding rod is fed into the weld by the other hand. The manual activities in carbon-arc welding are seen to be quite similar to those in gas welding. In neither case is it necessary for the material of the welding rod to traverse the arc.[1]

In metal-arc welding, we find a fundamental difference in this latter respect, since the arc, instead of having a graphite electrode for one terminal of the circuit, is established between a steel welding rod (or

[1] Both for carbon-arc welding and gas welding, the edges of the parts to be joined sometimes may be so designed as to obviate the need for any additional material; in other words, no welding rod is necessary in such cases.

welding electrode) and the two steel parts requiring to be joined. The operator in Fig. 1 is employing metal-arc welding to build up an incorrectly machined crank-shaft journal. There is always 0.1 in. (2.5 mm.) or more between the end of the welding rod and the work, which is bridged by an electric arc. The form in which the steel exists during its passage from one end of the arc to the other is at present the subject of investigation by several independent experimenters. Their conclusions are awaited with interest. The material cannot pass as a continuous liquid stream, since then there could be no interruptions in the metallic circuit and hence there could be no

FIG. 1.—METALLIC-ARC WELDING.

arc. It can pass as a series of liquid drops, which can even momentarily short-circuit the arc, the duration of the short-circuit being too brief to be apparent to the operator or ordinary observer unaided by special apparatus. Or the drops can be so minute as to be incapable of effecting a short-circuit. If this should be the case, we can conceive of the metal passing as a stream of finely divided liquid. Still another possibility is that the steel may pass as a highly heated gas and condense on the opposite surfaces. It is suggested by physicists that, in its passage through the arc, the steel may undergo instantaneous transformations of which no human knowledge at present exists.

There would appear to be more of these complex possibilities in

metallic-arc welding than in gas welding or in carbon-arc welding. Nevertheless, it is metallic-arc welding that is at present proving attractive to engineers. It is too early to say whether this widespread tendency toward metallic-arc welding is based on sound premises or whether there ultimately may not be a reaction (for certain kinds of work) back to carbon-arc welding. It may be that there has been undue precipitancy in the general movement from carbon-arc welding (which was the first to be developed) to metallic-arc welding.

Spot Welding

Spot welding, as developed for use in ship construction, consists in bringing into good contact, by hydraulic or pneumatic pressure, overlapping portions of the plates or parts requiring to be joined, and in sending through the spot of contact a sufficiently large current to heat the plates or parts at this point to a welding temperature. The weld is effected by the combination of pressure and heat.

Fig. 2.—Duplex spot welder.

Several large spot welders have been built. With one of these (which was an experimental machine), sufficient pressure and current were available to weld together three 1-in. plates. The usual construction of commercial spot welders for use in shipbuilding is similar in general appearance to so-called Bull riveters. The largest spot welder yet built for actual use in ship fabrication has a 6-ft. (1.8 m.) gap. This outfit is a large stationary machine to which the steel plates and shapes must be brought. It is planned that bulkheads, frames, floors, and other parts shall be constructed with it and shall then be transported by cranes to their places in the ship. This 6-ft. gap machine is designed with capa-

city to weld two ¾-in. plates. It provides a pneumatic pressure of 60,000 lb. and a current of 50,000 amp. and welds simultaneously two spots, each about 1½ in. diameter, in about 30 sec. With less current a longer time is required, and vice-versa. This particular spot welder, which is shown in Fig. 2, was built by the General Electric Co. and is known as a Duplex welder. This name is due to the feature that two spots are simultaneously welded, the current crossing the plate in one direction between two electrodes and then back again between two other electrodes. Two transformers, one located on each side of the plate, are comprised in the outfit. The arrangement is indicated diagrammatically in Fig. 3, in which *A* represents the two primaries, *B* represents the two secondaries (which, in the actual construction, have only one

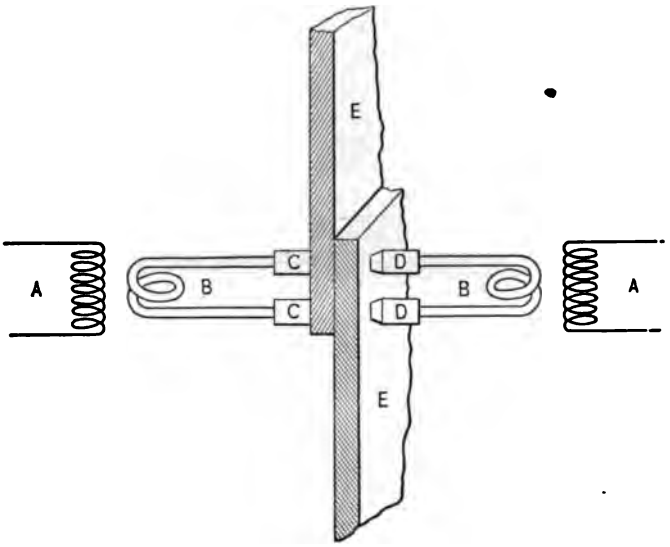

FIG. 3.—WIRING DIAGRAM OF DUPLEX SPOT WELDER.

turn each), *C* and *D* represent the electrodes between which the current flows and between which the pressure is exerted, *E* represents the two plates to be joined.

The chief object of this duplex feature is to eliminate the large reactance drop of a conducting loop about 6 ft. long and 1 ft. wide when traversed by 50,000 amp. of 60-cycle current. This amounts to approximately 25 volts. Mr. J. M. Weed, the designer of the machine, reports that the presence in the gap of the plates to be welded only decreased the current 10 per cent. This 6-ft. gap machine weighs 6 tons.

In some other large spot welders of somewhat reduced size and capacity, the duplex feature is not employed and only one spot is welded at each application of the current. A portable welder of this type, built by the General Electric Co. and having a 27-in. (68.5 cm.) gap, is shown in

Fig. 4. This machine weighs only 2800 lb. (1271kg.). In this case only one transformer is employed, and the circuit connections are those shown in Fig. 5, in which: *A* represents the primary of the transformer, *B* represents its secondary, *C* and *D* represent the electrodes, *E* and *F* represent the two plates to be welded. In all spot welders for welding thick plates, the electrodes are water-cooled.[2]

When, in the spring of 1918, Prof. C. A. Adams, of the Welding Committee of the Emergency Fleet Corporation, appointed the members of the Welding Research Sub-committee, we found ourselves facing a task of great interest and importance and of enormous magnitude. It was desired that our investigations should be directed chiefly to the

FIG. 4.—PORTABLE SPOT WELDER WITH 27-IN. GAP.

application of welding in the construction of the hulls of merchant ships. Much interior work on ships was already being performed very successfully with fusion welding and it appeared to be strongly indicated that the time was ripe for the extension of the application to the hulls of ships, with the prospect of producing work not fully equal (and probably superior) to that obtained by riveting, but also distinctly quicker and cheaper. At that time a welded barge was already nearly completed in England. The Welding Research Sub-committee ascertained that several American railways had for some time employed fusion welding extensively in routine repairs of locomotives and that possibly a

[2] Excellent discussions of the subject of spot welding and descriptions of several spot welders built for use in ship construction are given in the four following papers in the *General Electric Review*, December, 1918: Research in Spot Welding of Heavy Plates by W. L. Merrill, 919; Spot Welding and Some of its Applications to Ship Construction by H. A. Winne, 923; An Electrically Welded Freight Car by Jos. A. Osborne, 912; Some Recent Developments in Machines for Electric Spot Welding as a Substitute for Riveting by J. M. Weed, 928.

couple of thousand arc welders were at that time employed by American railways. The extensive and successful use of fusion welding for locomotive repairs, in itself, constituted strong evidence of the ability of such welds to withstand vibration and shock in addition to their proved excellence with respect to tensile strength.

Any doubts entertained by the Committee related chiefly to the question of which, of many ways in which it had been demonstrated that good fusion welding could be done, was the best way. Furthermore, as regards such mild-steel plates as are employed in the construction of merchant ships, it was soon demonstrated that while sound and quite ductile welds could be depended upon for plates not over ½-in. thick, there was less certainty of good results with plates of greater thickness. But at that time there was no general recognition of the most suitable current to be employed for welding. It was rare to find more than 150 amp. used, even for the heaviest work, and as low as 100 to 125 amp. was found to be frequently employed for welding plates of ½-in.

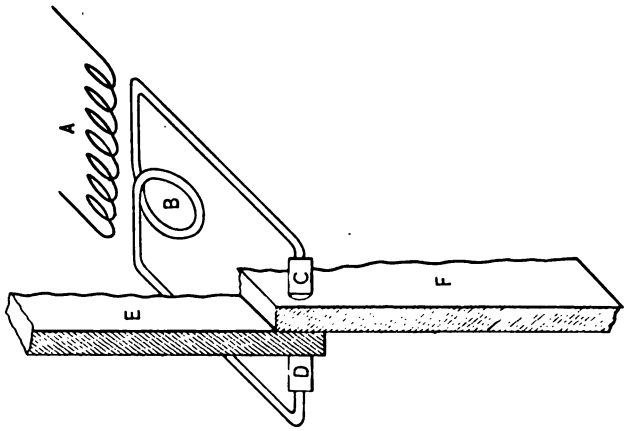

FIG. 5.—WIRING DIAGRAM OF PORTABLE SPOT WELDER WITH 27-IN. GAP.

thickness. It now has been quite conclusively shown that stronger and more ductile welds of ½-in. plates are obtained by using at least 200 amp. The author believes that fully 300 amp. should be used for butt-welding ¾-in. plates and at least 400 amp. for 1-in. plates. These currents are about twice as great as have heretofore usually been employed. In view of this subsequent experience, it is clear that the disappointing lack of strength and ductility in certain welds of thick plates made nearly a year ago was the result of using such small currents.

It would be easy to yield to the temptation to enter discursively upon comments and opinions regarding the many points on which experienced welding specialists hold widely diverging opinions. All these specialists are producing thoroughly reliable work, but this is not saying that they

are all producing as good work as could be produced under the most appropriate conditions for each case. Indeed, the author's observations lead him to the conclusion that while excellent arc welding is being done on a wide scale, there is room for improvement over the present average quality, which, so far as it can be expressed by a sort of resultant of such physical characteristics as bending and torsion tests, tensile strength and elongation at fracture, may be assessed as amounting to at least 25 per cent.

The author has attempted to make a list of some of the points which are the subject of discussion, and while usually not going at much length into the questions to make reference, in some instances, to the views and evidence on each side of a question. It has been the author's thought that such a summarized presentation might constitute the foundation for an instructive discussion.

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1. Bare and Covered Welding Wire

The kinds of electrodes advocated and actually used range all the way from the cheapest fence wire, costing but a few cents per pound (say at present prices, 10c. per lb.) up to carefully treated and covered electrodes. Some makes of covered electrodes are, however, obtainable at reasonable cost (such as, at present prices, 20c. to 25c. per lb.). One type of electrode must yield results uniformly superior to those obtained with another type in order to afford economic justification for a five-times greater price.

2. Salvaging Welding Wire

As to bare electrodes, it is generally considered that uniformity is very essential. Heretofore when faults have been found in the welding wire, it has been considered necessary to reject such wire. Some people now claim, however, that by merely dipping the electrode wire in suitable material it may be salvaged. Thus in the Welding Committee's Specification for Electrode Wire (given on p. 580) occurs a note to the effect that "If electrodes to the above Specification sputter or flow unevenly, they may be dipped in milk of lime (whitewash) before welding. This dipping may be done in quantity on stock on hand and allowed to dry, or the welder may keep a pot of solution on hand into which the electrode may be dipped immediately before welding." This method of salvaging electrode wire was developed by the Schenectady Research Laboratory of the General Electric Company.

3. Preferable Covering for Welding Wire

With regard to covered electrodes, while some persons claim that a thin covering obtained by dipping accomplishes the desired purpose, others contend that it is desirable to provide a thick covering of appropriate material, which, in turn, is suitably impregnated. Moreover, even for covered electrodes, the usual belief is that the greatest care should be given to the composition and quality of the welding wire to which the covering is applied. In other words, it is not generally held that the use of salvaged inferior wire will give the best quality of welds. It is important that the covering shall be consumed at a definite rate as compared with the rate of consumption of the enclosed welding wire. A result is that any particular gage of covered welding wire must be used within rather close current limits.

For overhead welding one firm exploiting covered electrodes supplies a special (and additionally high-priced) grade in which the covering is impregnated with a more viscous material than is used for the electrodes that the firm supplies for other welding operations.

4. Preferable Composition for Bare Welding Wire

There is a great diversity of practice as to the preferred composition of bare electrodes suitable for welding mild steel plates. As instances of extremes it may be said that amongst widely used electrodes, while one type consists of almost pure iron, other types have nearly 0.2 per cent. of carbon and 0.5 per cent. of manganese, and still other types run very much higher than this in manganese. This is quite aside from the subject of special compositions for welding high-carbon steel and for welding cast iron. It is anticipated that quantitative measurements will indi-

Table 1.—*Composition of Welding Electrodes for Metal Arc Welding*

Trade Designation of Electrode	Carbon	Manganese	Phosphorus	Sulfur	Silicon	Remarks
Page Steel & Wire Co.						
Armco	0.01	0.025	0.005	0.025	0.005	
Wilson Welders & Metals Co.						
Grade, No. 6	0.15 to 0.23	0.60 to 0.75	less than 0.04	less than 0.04		Also 0.25 per cent copper.
Grade, No. 9	0.30 to 0.40	about 1.00	less than 0.04	less than 0.04		
Grade, No. 8	0.17 to 0.22	0.30 to 0.45	less than 0.04	less than 0.04		
Grade, No. 17	0.10	0.30 to 0.45	0.06	0.06		
Quasi Arc Co	0.08 to 0.12	0.45 to 0.55	0.00 to 0.06	0.00 to 0.06	0.05 to 0.08	Flux covering of blue asbestos fiber (Crocidolite) enclosing percentage of aluminum or other metal in form of fine wire capable of giving strong reducing action.
Roebling Co	0.16	0.56	0.032	0.024	0.016	
Tonean wire	0.10	0.16	0.01	0.046	trace	
Electric Arc Cutting & Welding Co.	0.25	0.30	0.05	0.05	0.05	
Siemund Wenzel Co	0.10 and under	0.30 to 0.50	0.05 and under	0.05 and under	trace	
Norway-iron Wire	0.05	0.02	0.025	0.007	0.08	
Double Arc Co., of England	0.085	0.35	.....	0.054	0.108	Flux covered.
T. Scott Anderson Co., of England.	0.057	0.32	.....	0.026	0.014	Flux covered.
E. A. Jones & Co., of England	0.22	0.25	0.001	0.026	0.024	Nickel-plated and flux covered.
Engineering and Equipment Co., of England	0.12	0.51	.....	0.08	0.016	Flux covered.
Central Steel & Wire Co.						
Swedox	0.05	0.18	0.04			
The Spencer Wire Co., basic open-hearth steel electrode	0.06	0.12 to 0.20	0.013 and under	0.03 and under		

cate superiority in tensile strength for some compositions and superiority in ductility for other compositions. Mr. R. E. Wagner has exhibited some very ductile welds made with electrodes containing small percentages of magnesium and of boron sub-oxide.

It is only within the last few months that there have been available any specifications for use in establishing the merits of welding wire. These are now available in the Welding Committee's specification setting forth a "standard procedure for testing welding electrodes." This specification, which is given in Appendix A to this paper, was prepared by the Welding Research Sub-committee in collaboration with Prof. H. L. Whittemore representing the Bureau of Standards and with representatives of manufacturers of welding electrodes. In Table 1 are given the compositions of various electrodes in current use. The American Steel & Wire Co. requested the omission from this table of any analyses of electrodes it has furnished for arc welding for the reason that material has been supplied to a large number of users, varying considerably in analyses in accordance with the ideas of the purchasers. There does not as yet seem to be an agreement as to the most advantageous chemical composition for electrodes, and the company is not prepared either from observations of the results obtained by its customers or from its own experimental work to make a definite recommendation.

The Welding Committee has issued the following Specification for electrode wire for electric welding. The specification was prepared under the immediate direction of Mr. Hermann Lemp.

SPECIFICATION FOR ELECTRODE WIRE FOR ELECTRIC WELDING IN CONNECTION WITH MILD STEEL

Welding Committee Emergency Fleet Corporation
REVISED TO DECEMBER 20, 1918

(Note:—This wire may or may not be covered)

1. *Chemical Composition*

	NOT OVER, PER CENT.
Carbon	0.18
Manganese	0.55
Phosphorus	0.05
Sulfur	9.95
Silicon	0.08

2. *Sizes and Weights*

Diameter, in Mils	Diameter, in Fractions of an Inch	Pounds per 100 Ft.	Feet per 100 Lb.
125	1/8	4.16	2400
156	5/32	6.51	1535
188	3/16	9.37	1066

(Allowable tole ance 6 mils plus or minus)

3. *Material.*—The material from which the wire is manufactured shall be made by any approved process. Material made by puddling process not permitted.

4. *Physical Properties.*—Wire to be of uniform homogeneous structure, free from oxides, pipes, seams, etc., as proved by photomicrographs.

5. *Workmanship and Finish.*—(*a*) Electric welding wire shall be of the quality and finish known as the "Bright Hard" or "Bright Soft" finish—"Black Annealed" or "Bright Annealed" wire shall not be supplied.

(*b*) The surface shall be free from rust, oil or grease; a slight amount due to lubrication during last drawing is permissible.

6. *Tests.*—Electrodes must, before shipment or after delivery, show good commercial weldability when tested by an experienced arc welder. The electrode material shall flow smoothly in relatively small particles through the arc without any detrimental phenomena.

Note.—If electrodes to above specifications sputter or flow unevenly, they may be dipped in milk of lime (whitewash) before welding. This dipping may be done in quantity on stock on hand and allowed to dry, or welder may keep a pot of solution on hand into which the electrode is dipped immediately before welding.

7. *Delivery, Packing, and Shipping.*—Electrodes shall be furnished in straight lengths of either 14 in. or 28 in., put up in bundles of 50 lb. or 100 lb. as ordered. Each bundle shall be wrapped in heavy paper securely wired and marked on one end showing diameter, in mils, trade name, and grade of wire.

5. Composition of Metal Deposited in Weld

A few analyses have been made of chemical compositions of the metal deposited in the weld. Results of the analyses of four sets of electrodes before and after the metal was deposited are quoted below from the Westinghouse chapter in Major Caldwell's report. To these results are added analyses of Toncan wire as supplied to the author by Mr. R. E. Wagner.

Analyses of Electrode—Per Cent. of Impurities

	Carbon	Manganese	Phosphorus	Sulfur	Silicon
Roebling	0.16	0.56	0.032	0.024	0.016
Norway	0.049	0.021	0.025	0.007	0.08
C. R. S.	0.11	0.72	0.097	0.123	0.011
H. R. S.	0.13 to 0.17	0.50	0.012	0.045	0.011
Toncan	0.10	0.16	0.010	0.046	trace

Analyses of Deposited Metal—Per Cent. of Impurities

	Carbon	Manganese	Phosphorus	Sulfur	Silicon
Roebling	0.05	0.18	0.031	0.036	0.011
Norway	0.05	0.018	0.020	0.072	0.011
C. R. S.	0.05	0.11	0.086	0.072	0.011
H. R. S.	0.14	0.14	0.012	0.039	0.011
Toncan	0.042	0.081	0.019	0.026	0.000

It is notable that most of the carbon and manganese is burned out in traversing the arc.

6. Polarity

For carbon-arc welding, the standard practice is to connect the graphite electrode to the negative terminal. Mr. Wagner states as his experience that it is very difficult to weld with the carbon arc when the polarity of the carbon is positive. He states it to be almost impossible to direct the heat to the point desired and the welding qualities of the arc under this condition are very poor. He concludes: "Our experience has taught us that it is next to impossible to weld with a carbon arc unless the work is positive and the electrode negative."

For metallic-arc welding with bare wire, the electrode is usually connected to the negative terminal, but instances occur of bare welding wire that works best when the opposite polarity is employed. Also, for some particular sizes and sorts of welds, best results are sometimes obtained by a reversal of the polarity. With electrodes heavily covered with flux, the positive terminal is almost always connected to the electrode. Plenty of more or less plausible reasons for these differences have been offered on various occasions but none of them are very satisfactory. Among other considerations, the practicability of arc welding from an alternating-current circuit and of overhead welding have to be taken into account in judging some of these explanations. As yet, we have no satisfactory hypothesis as to what goes on in the welding arc.

7. Direct Current Versus Alternating Current for Arc Welding

While up to rather recently it had usually been contended that arc welding required a direct-current supply, there are now many advocates of alternating current. Mr. E. H. Jones in the course of the discussion of Major James Caldwell's paper entitled "Notes on Welding Systems" read on Jan. 22, 1918, before the Institution of Engineers and Shipbuilders in Scotland, stated that he would like to draw attention to the undoubted merits of alternating current for arc welding. For some reason, which he was unable to fathom, the general impression was that direct current was superior to alternating for arc welding, but as a matter of fact he found that alternating current was far superior to direct current, and he would recommend the use of alternating current on every possible occasion. Apart entirely from the capital outlay needed, which was vastly higher in the case of direct current, the control of the current was much easier to effect. . . . He estimated that the amount of current which would be necessary to feed twenty operators with direct current would suffice to feed twenty-eight with alternating current.

Mr. R. E. Wagner's experience is as follows: Electric welding may be done with alternating current as well as with direct current. It is a little more difficult to hold the arc, but this simply resolves itself into a matter of practice. Men who have been regularly doing arc welding with direct current very quickly learn how to handle the alternating current arc.

At present there appears to be no agreement as to the applicability of alternating current to carbon-arc welding.

8. Periodicity for Alternating-current Arc Welding

Among the advocates of the use of alternating current, there is no agreement with reference to the periodicity. Although it is generally maintained that arc welding is only thoroughly practicable with as high a periodicity as 50 or 60 cycles per second, it has been said that the use of 25 cycles, or less, is equally satisfactory. In October, 1918, Mr. R. E. Wagner reported to the Welding Research Sub-committee, that at the Pittsfield Works of the General Electric Co. he had found that alternating current for arc welding could be used with a frequency as low as 12½ cycles and as high as 500 cycles. Mr. Wagner states that while there is no difficulty at either of these extreme periodicities, the arc is more readily held at 500 cycles than at 12½ cycles.

9. Bare or Covered Electrodes for Alternating-current Arc Welding

While some persons maintain that arc welding with alternating current is only at its best when flux-covered electrodes are used, it appears to have been conclusively demonstrated by others that excellent results are being obtained under commercial conditions with bare electrodes and an alternating-current supply. A novice can more quickly learn to weld from an alternating-current supply if he employs flux-covered electrodes. But if he can ultimately learn to weld just as rapidly and successfully with bare electrodes, the difficulties in the initial stages of his education should not be regarded as being of much consequence. Mr. Wagner finds that when welding with alternating current, "manipulation may be simplified in many cases by treating the electrode with a thin coating of ordinary lime."

10. Relative Speeds of Alternating-current and Direct-current Arc Welding

Some contend that alternating-current welding is slower than direct-current. As an instance of a diametrically opposite experience, the

following recently received record of test of a certain electrode may be quoted:

"Its operation on 140 amp., 115 volts alternating current is very good. It also works satisfactorily on 130 amp., 75 volts direct current, but the metal flows more slowly on direct current than on alternating current."

The record concerning another type of electrode tested on the same occasion, reads as follows:

"This electrode was tried on $\frac{5}{16}$-in. plate, 120 amp., direct current, 75 volts. Its operation is satisfactory. It also works satisfactorily at 140 amp., alternating current, 115 volts, but its operation on alternating current is not quite as good as on direct current."

Mr. R. E. Wagner, who has had much to do with the development of both kinds of welding, states that "on the average the speed of welding with alternating current and direct current are about the same. We have had cases where alternating current is faster and vice versa."

11. Comparative Quality of Arc Welds Made With Alternating Current and with Direct Current

It is contended that the greater difficulty of maintaining an alternating-current arc (involving the necessity of acquiring the skill to hold a very short arc), entails as a consequence that an alternating-current weld is of superior quality. An experienced observer reports his experience as follows:

"Tests made have demonstrated conclusively that it is possible to do as good, or perhaps better, welding with alternating current as with direct current. No very decided difference has been noticed between welds made with alternating current and those made with direct current, but the welder who did most of the alternating-current welding says that, in his opinion, the alternating-current welds are better than the direct-current welds. This same opinion has been expressed by the machinist who made repairs on a small tank welded with alternating current. He said that the weld metal was better, more dense, and had fewer blowholes than a direct-current weld."

But an opinion from another authority of great experience in arc welding is as follows:

"As regards the strength of an alternating-current weld there is not the slightest doubt that a greater strength can be gotten on a test piece if that is all the work the man is going to do for some time. The facts, however, are that as the man's hand becomes fatigued in holding the alternating-current arc, his consequent breaking of the arc becomes more frequent, which means less strength in the weld, because every time the arc is broken, a bad spot is left in the weld."

12. Power Factor for Alternating-current Arc Welding

A view presented with considerable persistency is that the low power-factor associated with alternating-current welding leads to capital and operating costs off-setting any advantages. One answer made is to the effect that since for ship welding on an extensive scale, motor-generators are required, this only affects the generator and its circuit and does not affect conditions as regards the motor or the circuit from which it is supplied.

13. Alternating-current Arc Welding is Essentially a Single-phase Load

Similar considerations are involved in regard to the necessity of providing for the characteristics of a single-phase load. It is well known that single-phase motors and generators are much heavier, more expensive, and less efficient than polyphase motors and generators. With 30 or 40 arc welding outfits distributed fairly evenly on the different phases of a polyphase system, the load would be sufficiently balanced to be satisfactory, but this would correspond to an unusually large welding installation. In most cases it will be necessary to arrange for the welding to constitute a single-phase load and to make adequate provision to obtain satisfactory service with this condition.

Regarding the possibility of improving the power factor, Mr. W. S. Moody makes the following very suggestive statement:[1]

"Where a number of arcs are to be used within a reasonable distance of each other, the series system may be used. In this arrangement the secondary of an ordinary constant-current transformer supplies current to the primary of all the welding transformers in series. The individual transformers insulate the welding apparatus from the series circuit and transform from the series current to current of proper value for the arc. In this case the inherent reactance of the series transformer is low, but other features of the design are the same as those discussed above. The power factor of such a system can be safely made much higher than where individual arcs are operated in multiples from constant potential circuits."

14. Spot Welding is a Single-phase Load

Mr. J. M. Weed, who has had a great deal of experience with large spot welders, has kindly written the following paragraph on this subject:

"For welding plates from $\frac{3}{8}$ to $\frac{3}{4}$ in. in thickness, the single phase

[1] *General Electric Review* (December, 1918) . 937

currents required would be from 30,000 to 50,000 amp. and the kilovolt-amperes required at 60 cycles would range between 300 and 900 at power factors of from 0.35 to 0.50. These low power factors, combined with the fact that this load would be for short periods at very frequent intervals, would make it decidedly undesirable from the central station standpoint. The condition would be much improved at 25 cycles, as the same machine would operate equally as well at 25 cycles as at 60 cycles, with about one-half the kilovolt-amperes and about double the power factor. The intervals of operation would, however, be the same as for 60 cycles. If, however, a motor-generator set, with suitable flywheel attached, is provided for operating these machines, these disadvantages are all practically eliminated, this arrangement being such that the motor stores up energy in the flywheel during the interval of no load, the flywheel supplying a large part of the energy during the period of welding. By this means, for instance, a single-phase load of 900 kilovolt-amperes at 0.50 power factor for 30 sec. periods and with intervals of 1½ min. between periods would be converted to a practically continuous three-phase load of approximately 200 kilovolt-amperes at about 0.85 power factor."

15. Ductility of Arc and Gas Welds

Attention has been pertinaciously drawn to results of a very few tests which have appeared to indicate that metallic-arc welds are inherently utterly deficient in ductility, yet the Committee has had also before it the results of many well-authenticated tests of ductile metallic-arc welds. It has been claimed that gas welds are more ductile. On this matter Mr. R. E. Wagner writes:

"At several meetings of the Welding Committee, special stress has been brought to bear on the bending qualities of acetylene and arc welds. We have done some experimenting with average acetylene and arc welders, and our impression is that the acetylene and arc welds are in the same class with respect to bending. I submit herewith (see Fig. 6) a photograph showing comparative bends in acetylene and arc-welded joints, both welds were taken from ½-in. plate and both samples were bent under the same conditions, that is, the sharp edge of an angle iron was placed along the weld and pressure applied to the angle iron to make a sharp bend. These, I think, are average comparative results. * * * As far as our experiments are concerned, we feel, as regards physical characteristics, that acetylene and arc welds are in the same class."

16. Respective Fields of Gas and Arc Welding

On this subject, under date of Oct. 22, 1918, Mr. R. P. Jackson, of the Westinghouse Electric & Mfg. Co., reports to the Welding Research Sub-committee as follows:

"With reference to the comparative uses or fields of gas and arc welding which came up at the last meeting, it was suggested that some of us express our opinions on the matter based on our experience with both kinds of welding. In general, we have found gas welding to be more satisfactory for thin material, say ⅛ in. and under, and for general repair work, particularly where various kinds of steel and cast iron are involved. For example, if repairs have to be made on broken machinery, lugs rebuilt, pieces attached to high-carbon steel and work of this character, the gas-welding methods are superior and the extra cost not ordinarily prohibitive. When it comes, however, to depositing a large amount of metal and welding up structural steel or plates of ¼ in. thickness and upward, the results obtained by the ordinary direct-current arc with the metal electrode are at least equal to the gas welding work and certainly cheaper. In general, too, the finish of gas welding is more regular and better look-

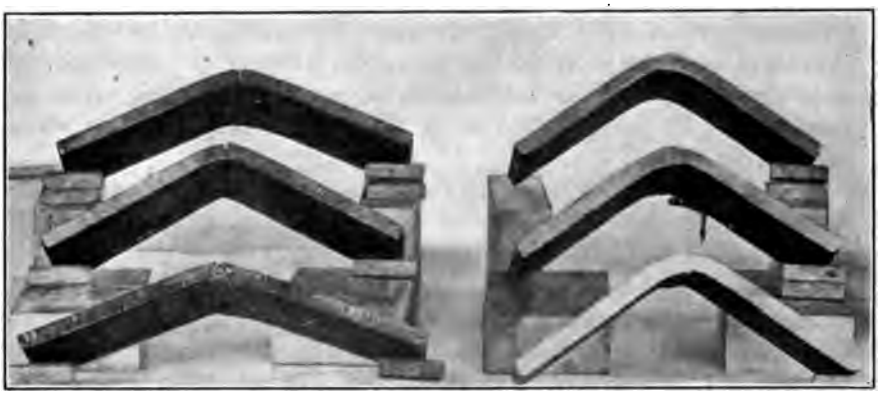


FIG. 6.—GAS AND ELECTRIC BENT SAMPLES. GAS WELDS AT LEFT AND ELECTRIC WELDS AT RIGHT.

ing and where that is a consideration gas welding may be preferred. In fact, in the Westinghouse factory at East Pittsburgh, there has been considerably more gas welding than electric, but the use of electric arc welding is increasing, not so much in displacing gas as in displacing riveting."

A view taken from a gas-welding publication is as follows:

"The arc process is chiefly used for filling up blowholes in large steel or iron castings and building up worn surfaces that are not to be machined. With this process the results obtained are somewhat uncertain and it is generally conceded, apart from the vital question of cost, that fusion produced by the burning of gases is to be preferred to the electric process. Welds made by the electric process are sometimes rough, hard, brittle and unworkable—in most cases this is highly objectionable, but not always so. With any fusion method of welding, annealing of the metal adjacent to the weld is desirable. It is impossible to do this

annealing with an electric welder, but with gas welding the blowpipe flame can be used for heating the metal surrounding the welded part, and also for heating metal away from the weld, so as to counteract any strains that may be set up in the piece as the weld cools off. There are certain classes of work for which electric welding is the most suitable system, and, on the other hand, there are many classes of work where it would be most impractical, and which can be done satisfactorily only with gas welding. For general workshop use, a gas welding outfit is far better, not only because of its greater economy in installation and operation, but also because of its wider range of usefulness."

In the absence of any experience to the contrary, this latter view appears fairly plausible, and it is natural that it should have received wide acceptance. But experience in arc welding has gradually been acquired and it controverts the correctness of the view. Unfortunately, the experimental data available on the subject of gas welding are surprisingly meager. The Welding Research Sub-committee has concluded that there are practically no test data from which it can draw any safe generalizations as to the mechanical characteristics of gas welds, and that it will be necessary to make its own investigations to obtain suitable data.

Gas welding was an established art before there was any large amount of electric welding. This was still the state of affairs in England until shortly before the author was there in the autumn of 1917. But the war conditions had occasioned in England such a shortage of supplies of oxygen and carbide that the Government, as a war measure, practically forced the wide substitution of arc welding for gas welding. The British Government, in entering upon this policy, had relatively little concern as to the comparative merits of the two methods except in so far as that any merit or advantages found to be associated with arc welding would naturally assist in bringing about its substitution for gas welding.

It was, however, with considerable surprise that it was ascertained that the true economic field for arc welding, as compared with gas welding, was a very wide one, and that, simply due to inertia and tradition, engineers had been continuing in the contrary belief. Major James Caldwell, of the Admiralty Controller's Department, had great responsibilities in this task of substituting arc welding as rapidly and generally as possible. Major Caldwell provided the author with the results of his investigations into the relative costs of gas and electric-arc welding. These results, which correspond to conditions in December, 1917, are set forth in Table 2, which shows that electric-arc welding was found to be a faster process for all thicknesses of steel. The British Admiralty results furthermore indicate the economic field for the two methods; they are in favor of gas welding for thin plates and of electric-arc welding for thick plates. But

TABLE 2.—*Comparative Costs of Welding by Oxyacetylene and Electric Arc*

Thickness of Metal, Inch	Oxyacetylene					Electric Arc					Oxyacetylene		Electric Arc	
	Gas per Hour		Cost for Gas per Foot Run, Pence	Iron Wire for Filling, Pence	Labor per Foot Run	Power per Foot Run			Labor per Foot Run, Pence	Cost of Electrodes per Foot Run, Pence	Feet Run per Hour	Total Cost per Foot Run, Pence	Feet Run per Hour	Total Cost per Foot Run, Pence
	Oxygen, Cubic Feet	Acetylene, Cubic Feet				Volts	Amp.	Kw.						
1/16	3	2.0	0.116	0.131	0.40	100	30	0.075	0.3	12 in. No. 12 = 1.2	30	0.647	40.0	1.575
1/8	9	6.3	0.77	0.196	0.856	100	75	0.312	0.5	18 in. No. 12 = 1.8	14	1.822	24.0	2.612
3/16	13	9.0	1.72	0.262	1.33	100	100	0.50	0.6	24 in. No. 10 = 2.64	9	3.312	20.0	3.74
1/4	17	13.0	2.70	0.250	1.50	100	140	0.70	0.6	22 in. No. 8 = 3.6	8	4.45	20.0	4.90
5/16	27	16.0	4.21	0.327	1.714	100	110	1.10	1.2	12 in. No. 10 18 in. No. 8 = 4.02	7	6.25	10.0	6.32
3/8	34	24.0	6.82	0.458	2.00	100	120	1.61	1.61	18 in. No. 10 24 in. No. 8 = 5.58	6	9.278	7.4	8.802
7/16	41	29.0	9.90	0.655	2.40	100	120	2.00	2.0	24 in. No. 10 36 in. No. 8 = 8.04	5	12.955	6.0	12.04
1/2	48	34.0	14.50	0.786	3.00	100	120	2.40	2.4	30 in. No. 10 42 in. No. 8 = 9.6	4	18.286	5.0	14.40

Labor taken at 1 shilling per hour.
Oxygen taken at 1/2 pence per cu. ft.
Acetylene taken at 1 pence per cu. ft.
Iron for filling taken at 0.131 pence per ft.
The above figures are based on the British Oxygen Co. standards.

Labor taken at 1 shilling per hour.
Current taken at 1 pence per B.O.T. unit.
Electrodes, No. 10, 1.32 pence per ft; No. 8, 1.8 pence per ft: No. 12, 1.2 pence per ft.

the comparison is based on the very high cost of electrodes set forth below:

Standard Wire Gage Number	Cost, in Cents per Foot	Feet per Pound of Contained Iron Wire	Cost, in Cents per Pound
8	3.6	15	54
10	2.6	23	60
12	2.4	35	84

By substituting a typical American price for labor and substituting the cost of bare electrodes, such as are used with entire success in America, for the cost of flux-covered electrodes, the results will show a lower cost for arc welding than for gas welding for all thicknesses above $\frac{1}{16}$-in. The question of the quality of the weld is another matter, but judging from the general reputation of the work of all sorts done by gas welding and by electric-arc welding, both are thoroughly reliable. No more exact comparison can be made till we have carried through to completion really elaborate tests of gas welds in order to permit of making a sound comparison with the large amount of research data already obtained with electric-arc welds.

In response to a request for his opinion as to the respective fields for gas and electric-arc welding, Mr. R. E. Wagner, of the Pittsfield Works of the General Electric Co., writes as follows: "The present well-tried field for electric-arc welding is confined entirely to welding plates and forms, and a great deal of work has been done on plates varying in thickness from $\frac{1}{16}$ in. to $\frac{3}{4}$ in. Up to $\frac{1}{8}$-in. plates, the cost of gas and electric welding is about the same; beyond this, the cost is in favor of the electric process. No difficulty is experienced in machining electric welds made with the metallic electrode. While it is recognized that the electric-welded-in material will not stand bending equal to that of the plate in which it is deposited, it is on the average equal to gas-deposited material in this respect."

17. Relative Ductility of Welds Made With Bare and Covered Electrodes

By some authorities, ductility is believed to be most readily obtained by employing flux-covered electrodes. On the other hand, the Committee has knowledge of several kinds of bare electrodes of various compositions which, in competent hands, make reasonably ductile welds.

18. Speed of Arc Welding

All sorts of values are given for the speed, in feet per hour, with which various types of joints can be welded. Operators making equally good

welds have widely varying degrees of proficiency as regards speed. Any quantitative statement must consequently be of so guarded a character as to be of relatively small use. In general, and within reasonable limits, the speed of welding will increase considerably when larger currents are employed. It appears reasonable to estimate that this increase in speed will probably be about 25 to 35 per cent. for high values of current. This increase is not directly proportional to the current employed because a greater proportion of time is taken to insert new electrodes and the operator is working under more strenuous conditions. Incidentally, the operator who employs the larger current will not only weld quicker but the weld will have also better strength and ductility.

On this point Mr. Wagner writes as follows: "I would not say that speed in arc welding was proportional to the current used. Up to a certain point ductility and strength improve with increased current, but when these conditions are met, we do not obtain the best speed due to increased heating zone and size of weld puddle. Speed may fall off when current is carried beyond certain points."

In a research made by Mr. William Spraragen for the Welding Research Sub-committee on several tons of ½-in. (12.7 mm.) ship plate, the average rate of welding was only 2 ft. (0.6 m.) per hr. Highly skilled welders were employed but they were required to do the best possible work, and the kinds of joints and the particular matters under comparison were very varied and often novel. However, it was found that about 1.9 lb. (0.8 kg.) of metal were deposited per hour when using a $\frac{5}{32}$-in. (3.9 mm.) bare electrode and with the plates in a flat position. The amount of electrodes used up was about 2.7 lb. per hr., of which approximately 16.5 per cent. was wasted as short ends and 13 per cent. burnt or vaporized, the remainder being deposited at the speed of 1.9 lb. per hr. mentioned above.

For a 12-ft. cubic tank of ½-in. steel welded at Pittsfield, the speed of welding was 3 ft. per hr. The weight of the steel in this tank was 16,000 lb. and the weight of electrode used up was 334 lb. of which 299 lb. was deposited in the welds. The total welding time was 165 hr. corresponding to using up electrodes at the rate of just 2 lb. per hr. The total length of weld was 501 ft., the weight of electrode used up per foot of weld thus being 0.60 lb. The design of this tank comprised eighteen different types of welded joint. Several different operators worked on this job and the average current per operator was 150 amperes.

For the British 125-ft. cross-channel barge for which the shell plating was composed of ¼-in. and $\frac{5}{16}$-in. thick plates, in Mr. H. Jasper Cox's paper read before the Society of Naval Architects on Nov. 15, 1918, and entitled "The Application of Electric Welding to Ship Construction" it is stated that: "After a few initial difficulties had been overcome, an

average speed of welding of 7 ft. per hr. was maintained, including overhead work which averaged from 3 to 6 ft. per hour."

In a report appearing on page 67 of the minutes and records of the Welding Research Sub-committee for June 28, 1918, Mr. O. A. Payne of the British Admiralty states: "A good welder could weld on about 1 lb. of metal in 1 hr. with the No. 10 Quasi-Arc electrode, using direct current at 100 volts. An electrode containing about 1½ oz. of metal is used up in about 3 min. but this rate cannot be kept up continuously."

The Quasi-Arc Co. publishes the following data for the speed of arc welding in flat position with butt joints, a 60° angle and a free distance of ⅛ in.: ⅛-in. plate, 30 ft. per hr.; ¼-in. plate, 18 ft. per hr.; ½-in. plate, 6 ft. per hr.; 1-in. plate, 1.3 ft. per hr. We cannot, however, reconcile the high speed of welding ½-in. plate, published by the Quasi-Arc Co. as 6 ft. per hr., with the report given by the British Admiralty that a good welder deposits 1 lb. of metal per hour with the Quasi-Arc electrode. If the rate given by the Quasi-Arc Co. is correct, it would mean that about 4 lb. of metal were deposited per hour. On this basis the rate must have been computed on the time taken to melt a single electrode and not the rate at which a welder could operate continuously, allowing for his endurance and for the time taken to insert fresh electrodes in the electrode holder and the time taken for cleaning the surface of each layer before commencing the next layer.

From his observations the author is of the opinion that a representative rate for a good welder lies about midway between the values given by Mr. Payne and by the Quasi-Arc Co., say for ½-in. plates some 2 lb. per hr. This, it will be observed, agrees with Mr. Spraragen's experience in welding up 6 tons of ½-in. ship plate with a dozen or more varieties of butt joint and Mr. Wagner's results with the 8-ton tank. Even this rate of 2 lb. per hr. is only actual time of welding operator after his plates are clamped in position. This preliminary work and the preparation of the edges, which is quite an undertaking, and requires other kinds of artisans, accounts for a large amount of time and should not be underestimated.

The practice heretofore customary of stating the speed of welding in feet per hour has led to endless confusion as it depends on type of joint, height of weld, and various details. A much better basis is to express the speed of welding in pounds of metal deposited per hour. Data for the pounds of metal deposited per hour are gradually becoming quite definite. The pounds per foot of weld of metal required to be deposited can be readily calculated from the drawings or specifications. With the further available knowledge of the average waste in electrode ends and from other causes, the required amount of electrode material for a given job can be estimated.

19. Suitable Current for Given Cases

For a given type of weld, for example, a double-vee weld in a ½-in. ship plate, it was found that while some operators employ as low as 100 amp., others work with over 150 amp. Some, in making such a weld, employ electrodes of only ⅛ in. diameter and others prefer electrodes of twice as great cross-section. For this particular size and design of weld, the Welding Research Sub-committee has had welds made with from 200 to 300 amp. The conclusion appears justified that the preferable current for such a weld is at least 200 amp. If the weld of the ½-in. plate is of the double-bevel type, about 50 amp. less current should be used for the bottom layer than is used for the second layer, if two layers are used. For ¾-in. plates, the most suitable welding current is 300 amp. This is twice the current heretofore usually employed for such a weld.

Mr. Wagner writes: "We have made a number of tests to determine the effect of varying current on the strength of the weld. Tests were made on a ½-in. plate with current values as follows: 80, 125, 150, 180, 220, 275, and 300 amp. These tests show improvement in the tensile strength and bending qualities of welds as the current increases. The speed of welding increases up to a certain point and then decreases."

20. Effect on Arc Welding of Voltage Employed

Mr. Wagner reports as follows: "We have made a number of tests to determine the influence of variable voltages on the strength and character of electric welds. The experiments were made welding ½-in. plate with 150 amp. held constant and voltage varying as follows: 40, 75, 100, 125, 150, 200 and 225 volts. This test demonstrates that there is no material difference in the tensile strength, bending qualities, or the appearance of the welded-in material. There is this advantage, however, in the higher voltage, that variations in the strength of the arc do not materially affect the value of the current. A curve-drawing ammeter was installed on the welding circuit which showed variations in current at 75 volts but at 150 volts the current curve was practically a straight line."

21. Preferable Size of Electrode

On certain railways, a single diameter of electrode is employed independently of the size or shape of the plates or parts being welded. The experience of other people leads them to use several different sizes of electrodes according to the size of the job and the type of joint. Present British practice appears to be to use such a size of electrode as to

have a current density of 4000 to 6000 amp. per sq. in. The investigations of the Welding Research Sub-committee are indicating that at least 10,000 to 12,000 amp. per sq. in. is suitable for electrodes of $\frac{1}{8}$ in. and $\frac{5}{32}$ in. diameter and well up toward 10,000 amp. per sq. in. for electrodes of $\frac{3}{16}$ in. and $\frac{1}{4}$ in. diameter.

22. Automatic Machinery for Arc Welding

Several firms are developing machinery for feeding the electrode automatically. Such machinery appears to be capable of making excellent welds at higher speeds than are attainable by hand feeding.

23. Carbon-arc Welding

With the advent of metallic-arc welding there has been a tendency to neglect the carbon-arc method. It is quite possible that this attitude is not justified for not only is there now a definite field where the carbon-arc method is advantageous but developments in the art may greatly extend its application. It is generally agreed that the carbon-arc method is not applicable to vertical and overhead welding, which is, of course, a serious handicap in ship-hull work. The opinion of competent observers (with, however, some emphatic dissenting views) appears to indicate that carbon-arc welding is not as reliable as metal-arc welding in ordinary welding, because: (*a*) Carbon is carried into the deposited material, thus reducing its ductility. (*b*) It is more difficult to obtain good fusion on account of overlapping of deposited metal on the original metal. (*c*) It is more difficult to manipulate and thus requires greater skill. (*d*) It is a much hotter arc, which means greater discomfort to the operator and therefore lower efficiency. (*e*) Greater cooling stresses are developed because larger areas of adjacent metal are heated. On the other hand, it is contended by some that carbon-arc welding can be developed to the point where these objections will no longer exist and thus gain the advantages of this method, the principal of which are: (*a*) No preparation of the abutting edges is necessary. (*b*) Greater rate of deposition of metal and therefore greater speed of welding, particularly in heavy work. (*c*) Probable greater adaptability to automatic welding. It should be stated that there is very general agreement as to the superiority of the carbon arc over the metallic arc for heavy work where strength is not so important, especially cast-iron welding and the filling of holes in iron and steel castings.

24. Preparation of Welding Edges

British practice permits the use of smaller angles when the edges of the plates are veed than does the American. If the smaller angles

give welds which are equally satisfactory in all respects, the decreased amount of electrode material required, the decreased consumption of electricity, and the increased speed are advantages not to be overlooked; but obviously the matter requires careful investigation. American practice which, up to recently, has been with a very wide angle, appears to have required the consumption of about twice as great a weight of electrode as British practice with the smaller angle. The urgent importance of determining whether the use of the smaller angle involves any sacrifice in quality is evident. There is already considerable basis for the belief that actually better results attend the employment of a smaller angle of bevel when a suitably large current is used. A shoulder in place of the heretofore commonly used sharp bottom edge of the bevel also constitutes a material gain not only in the saving in welding material, but also in the quality of the weld.

Mr. Wagner states that at the Pittsfield Works of the General Electric Co. they have long adopted the practice of using a 30° bevel for plate edgings and that they find it satisfactory for all thicknesses up to $\frac{3}{4}$ in. He states that this angle gives sufficient room for depositing the metal, reduces the time to weld and the amount of metal deposited.

In one of Mr. Spraragen's researches, various angles of bevel were used. Although the physical tests have not yet been made, Table 3 shows the time, amount of metal, and electricity consumed for these different angles of bevel. The "free distance" in each case was $\frac{1}{8}$ in. and the welding was done in a flat position with $\frac{5}{32}$-in. bare electrodes. In each case the weld had a length of 3 feet.

TABLE 3.—*Time, Metal, and Current used with Welds of Different Bevels*

	Angle of Bevel Used, in Degrees			
	15	30	45	60
Amperes	160	145	118	125
Weight of electrode used up, in pounds	2.56	3.83	4.63	6.63
Weight of metal deposited, in pounds	1.70	2.55	3.65	5.08
Weight of metal wasted, in pounds	0.86	1.28	0.98	1.55
Pounds deposited per hour	1.82	1.61	1.82	1.81
Feet welded per hour	3.22	1.90	1.50	1.07
Circuit kilowatts	9.91	9.00	7.68	8.25
Kilowatt-hours per foot of weld	3.10	4.70	5.10	7.70

25. QUALITY OF OVERHEAD ARC WELDING

The British Admiralty regards overhead welding as too inferior and too expensive to be employed when it can possibly be avoided. In America a large amount of overhead welding is done in railway shops

and it is claimed that it is simply a matter of training operators to the required degree of proficiency.

26. Number of Layers to be Employed

Good progress is being made in obtaining knowledge of the relative characteristics of welds made with different numbers of layers and of the most suitable current and the most suitable size and type of electrodes to employ for each layer. The tendency is toward the use of at least two layers for ½-in. plates, and three layers for ¾-in. plate.

27. Rigid vs. Non-rigid Methods of Welding

On this question it is more a matter of determining the conditions essential to obtaining good results with whichever of the two methods is most appropriate for each particular purpose. The term rigid is applied to the process of arc welding in which the two parts to be joined by welding are, prior to welding, held rigidly by bolting or clamping or by a series of preliminary tack-welds distributed at various points. The rigid plan is the most obvious for welding the hull plates of ships but its critics claim that the resultant joints are deficient in ductility due to the presence of internal stresses. It is considered that by suitably arranging the order of welding it is practicable to so distribute the heat as to avoid these stresses. At any rate there are many alternative orders of procedure in making welds by the rigid method and elaborate researches should be made to ascertain the procedure which will yield the best result.

The non-rigid method consists in placing at a slight angle to each other the two plates to be welded. As the welding operation progresses along the seam the angle gradually closes and when the weld is completed the width of the welded seam is equal throughout its extent. Such welds are generally considered to be very free from internal stresses, and hence more ductile.

28. Results of Different Lengths of Arc

The metallic-arc is much shorter than the carbon arc. As a result the metallic-arc weld has the advantage that there is less opportunity for oxygen and nitrogen to gain access to the weld and so far as relates to this feature the metallic-arc weld should be better. But with the carbon arc the added metal does not traverse the arc, the tip of the welding rod being held down close to the surface on which it is to be deposited. This may render the deposited material less subject to contamination in carbon-arc welding than in metallic-arc welding since it has not traversed the arc.

Coming to the exclusive consideration of metallic-arc welding, the

greater the welding current the less is the area represented by the cylindrical surface of the arc per pound of metal traversing the arc, and consequently the less should be the contamination by oxygen and nitrogen from the surrounding air. So far as this circumstance is concerned, the greater the welding current, for a given case, the greater should be the

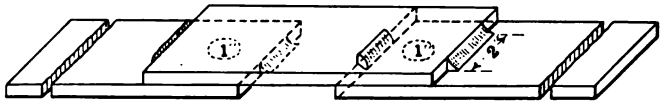


FIG. 7.—FILLET AND SPOT WELDED.

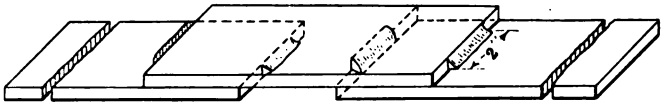


FIG. 8.—FILLET WELDED.

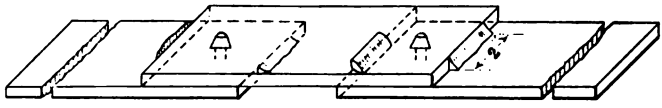


FIG. 9.—RIVETED AND FILLET WELDED.

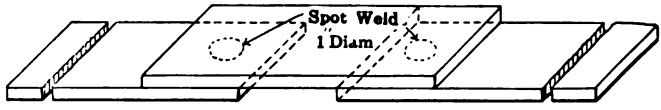


FIG. 10.—SPOT WELDED.

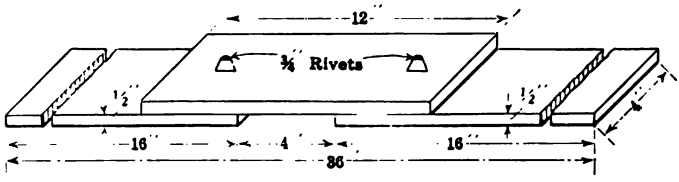


FIG. 11.—RIVETED JOINT.

ductility of the joint. On the other hand, it seems probable that even the most skilful operators will be unable to hold quite so short an arc with the larger current.

29. SPOT AND ARC WELDING

A good deal of progress is being made in America in the use of spot welding for the joining of thick plates. It is believed that spot welding has a great future as applied to shipbuilding and several large spot welders have been built for this purpose. In some of its applications, spot welding affords a method of preliminarily joining the hull plates,

after which the required additional strength is provided by arc welding. The Welding Research Sub-committee has already made some progress in comparing combined spot and arc welds and combined rivet and arc welds with riveted, spot-welded, and arc-welded joints. It is not a question in such an investigation of spot versus arc welding, but of spot and arc welding.

In the tests mentioned, the specimens were made up of the following combinations: Spot and fillet welded (Fig. 7) two samples. Fillet welded, made by welding fillets about 2 in. in length at the ends of the plates (Fig. 8) two samples. Riveted and fillet welded (Fig. 9) one sample. Spot welded, made by welding two spots approximately 1 in. in diameter, on the plates (Fig. 10) two samples. Riveted joint made by riveting a ½ in. by 4 in. by 12 in. plate with two plates ½ in. by 4 in. by 16 in., using two ¾-in. rivets and a 4-in. lap; one sample.

The results of the test show the comparative strength of the joints as follows:

	Ultimate Load, Pounds
Spot and fillet welded	50,350[a]
Fillet welded	37,000[a]
Riveted and fillet welded	35,000[b]
Spot welded	28,000[a]
Riveted joint	13,000[b]

Spot welds, as compared with arc-welded butt joints, have the disadvantage of the increased weight corresponding to the overlap.

30. Condition of Surfaces to be Welded

While for spot welding the surfaces may sometimes be too clean to obtain the best weld, this cannot be the case with fusion welding. The question of the extent to which it is practicable to go in freeing the surfaces from impurities prior to making the fusion weld is entirely a commercial one; the cleaner the surface, the better the weld. In spot welding it is desirable to have clean surfaces under the electrodes, but scale between the two plates is a positive advantage.

31. Preheating, Heat Treatment and Hammering While Cooling

Preheating, heat treatment, and hammering, as applied to fusion welding (both gas and electric) have been the subjects of research, but as yet nothing adequately comprehensive has been planned. It is very important that these deficiencies should be recognized and remedied.

[a] Average of tests on two samples. [b] Only one sample made.

32. Time Required to Train Welders

As to the training of welders, there are on the one hand those who are so ill-informed as to assert that good welding can be done after a very few weeks of training, while on the other hand, others assert that an operator can hardly acquire any considerable degree of skill in much less than a year. Still others advocate the substitution of automatic feeds and the complete elimination of dependence upon manual dexterity. Arc welding requiring more than 100 amp., in the author's opinion, is not a trade for women when the arc is manipulated by hand as distinguished from the use of automatic or semi-automatic machinery.

33. Need for Special Machines for Welding

A great variety of machinery for supplying and controlling the current for welding is on the market. Some of this machinery comprises elaborate mechanisms with which it is claimed that it would be very difficult for even a novice to make a bad weld. Some advocate the use of simple resistance to be inserted in series with the arc on any available circuit, and claim that any additional machinery is superfluous. The capital outlay for the equipment of a welder (at the point of consumption) when the first kind of equipment is used, may be $1000 or more, while in the second case, less than half of this sum is sufficient.

34. Technique of Testing Welds

The ideal weld should presumably be as strong and as durable as the metal of the members joined together. In other words, the section containing the weld should have the same chemical and physical characteristics as adjacent sections in the original metal. A weld is therefore measured by the degree of approximation to this condition, as determined by mechanical, chemical, and metallurgical tests of the parent metal, the welded joint, and the deposited material in the weld.

While during the last year the Welding Research Sub-committee has made great progress in establishing standard procedures for the mechanical testing of welds, much still remains to be done. Obviously, the procedure for testing the original metal should follow standard practice as recommended by the American Society for Testing Materials, but there is considerable difference of opinion and uncertainty as to just how and what mechanical tests should be made of the welded joint and of the deposited metal. For instance:

(*a*) Should all the usual observations be taken when making a tensile test of a welded joint? Obviously the strength of the union between the two pieces of metal should be determined but in view of the non-homogeneity of the specimen, does not a very different significance attach to yield point, elongation, and reduction of area? Where a series of welds

having the same ratio of deposited material to original metal is concerned, such data are undoubtedly important for comparison purposes but for evaluating a weld in terms of the original metal, questions are repeatedly being raised as to just what extent these data have value.

(*b*) Would not more reliable information as to the ductility of the weld be obtained if elongation and reduction of area measurements were made on specimens prepared from the deposited metal or from specimens cut lengthwise of the weld instead of crosswise?

(*c*) Similarly with the bending test, which is a test for ductility. There are some (including the author) who would make the bend with the axis of the mandrel normal to the weld instead of parallel thereto, which latter position is the one usually employed. It may be that both tests should be made; the normal position as test of the ductility of the deposited material and the parallel position as an additional test of the union between the deposited material and the original metal.

(*d*) How important are torsion tests and impact or shock tests in measuring welded joints?

(*e*) Fatigue tests of welded joints are generally conceded to be vital and the importance of obtaining reliable information as to how this test should be made probably transcends (at present at least) that attached to any other research in the field of fusion welding. The researches should be made: (1) With the Moore bending fatigue machine; (2) with rod samples rotated at high speed as employed by Lloyd's Register, in England; (3) with the Strohmenger torsion-fatigue machine; (4) with the Cammell-Laird bending fatigue apparatus; (5) by the Upton-Lewis test.

After the necessary research work has been done to solve these and other similar questions pertaining to the testing of welds, standard specifications for the testing procedure can be prepared which will be properly balanced between the cost of making the tests and the amount of testing necessary to insure a reliable estimate of the weld.

Conclusion

The extent of the field of application for fusion welding and spot welding is but little appreciated by engineers other than those who have been directly connected with welding developments. It is evident that this field is an enormous one, including as it does all structures where steel is employed, such as bridges, building structures, tanks of all kinds, railway rolling stock, and ships, in addition to numberless miscellaneous applications in industry in general. However, engineers associated with welding research should be on their guard, that their enthusiasm over this great field of application may not lead them into prematurely indorsing the use of fusion welding or spot welding in constructions where the consequences of failure involve serious menace to life and property, as may often be the case. For example, a particularly

important case is that of pressure vessels and especially large high-pressure containers. The success in one hundred installations will not excuse failure (accompanied possibly by fatalities), in the one hundred and first installation. It is the opinion among the best-informed engineers that, before fusion welding can advisedly be employed for large high-pressure vessels, much vigorous and elaborate research work should be carried out on the fatigue characteristics of fusion welds of long seams, and that this research work must comprise full-sized structures since the conditions cannot be reproduced in test samples. In fact, if the general acceptance of welding, particularly by inspection boards, underwriters, and classification societies, is to be accomplished in a reasonably short time, such extensive research work on a large scale is absolutely essential in order to demonstrate conclusively that welded joints are equal to or better than joints made by other methods. Obviously the development of the art could proceed along the lines of the usual order of evolution as in the cases of previous arts, but this would, as in those cases, involve the lapse of years. For structures subjected to less extreme stresses, such as the hulls of ships, the adequacy of fusion welding as a substitute for riveting is in process of being thoroughly demonstrated in actual practice in Great Britain. It is recognized that the hulls of ocean-going ships are exposed to very great stresses, nevertheless there is a clear distinction between the magnitude of those stresses and the stresses to which many large, high-pressure containers are subjected.

The author hopes this paper will aid in focusing attention on the vast importance of the welding art, particularly by occasioning discussion of the many problems in welding research, some of which have been mentioned. He cannot give adequate acknowledgment of his indebtedness to his many associates in the preparation of this paper. The most generous assistance has been given him on every hand. Mr. William Spraragen has extended much assistance in preparing data and in many useful ways and Mr. F. M. Farmer, chief engineer of the Electrical Testing Laboratories, has given very generously of his time.

APPENDIX A

Standard Procedure for Testing Welding Electrodes. Welding Committee, Emergency Fleet Corporation November, 1918

1. *Purpose.*—The purpose of this specification is to provide a standard procedure for testing welding electrodes for electric arc welding which are submitted for the information of, and with the view to the approval by, the Welding Committee of the Emergency Fleet Corporation.

This specification describes a test of electrodes and not a combina-

tion of an electrode and of an apparatus. The fact that the applicant is given the option of selecting the system with which the test is made does not make it a test of that system, any more than the employment of a particular welding operator could be said to make it a test of a welding operator. The system used in making these tests may or may not prove to be of importance. It is sought to minimize the influence of the individuality of the operator by requiring the test to include welds made by at least two operators.

The Welding Committee has two other Sub-committees into whose province falls the approving and certifying of operators and the approving and certifying of systems. The admission that these three tasks overlap is no reason for delaying the important task of proceeding at once with the standardizing of the testing of electrodes.

2. *General Conditions.*—(*a*) Each applicant shall provide at his factory, or elsewhere as he may elect, the necessary facilities and at least two operators for making the test welds prescribed by these specifications. Except when otherwise indicated in these specifications, the applicant shall select the apparatus and other conditions for making test welds which he considers most suitable for his electrodes.

(*b*) The test welds shall be made in the presence of an authorized representative of the Welding Committee of the Emergency Fleet Corporation who will be empowered to certify as to the compliance with the conditions prescribed in the specifications. Also representatives of Lloyd's Register of Shipping and the American Bureau of Shipping shall have the opportunity to witness the making of the test welds.

(*c*) Until otherwise declared, the authorized representative of the Welding Committee shall be the Electrical Testing Laboratories, 80th Street and East End Avenue, New York, N. Y.

(*d*) The Welding Committee shall assume charge of the completed test welds and have them tested by the authorized representative in accordance with these specifications.

(*e*) The authorized representative shall render to the Welding Committee a complete, detailed report of each test of an electrode.

(*f*) The cost of carrying out the prescribed tests shall be borne by the applicant.

3. *Sample Electrodes.*—About 100 lb. of each electrode to be tested, which will be known as the sample, shall be furnished without charge. It shall be accompanied by an affidavit giving the following information:

(*a*) The trade name under which the electrode is marketed together with certification that all electrodes bearing this trade name will be substantially the same as the sample submitted.

(*b*) The manufacturer of the complete electrode.

(*c*) The location of the factory in which the electrode was made.

(*d*) The manufacturer of the welding wire.

(*e*) The location of the mill in which the welding wire was produced.

(*f*) The chemical analysis of the welding wire.

The authorized representative shall retain the remainder of the sample for reference purposes.

4. *Plate Material.*—For these tests ½-in. ship plate shall be used. This material shall comply with the "*Standard Specifications for Structural Steel for Ships*" as adopted by the American Society for Testing Materials, serial designation A 12–16 (see page 98, A. S. T. M. Standards, 1918).

An abstract of this specification is as follows:

1. Open-hearth steel.
2. Phosphorus { acid steel, not over 0.06 per cent. / basic steel, not over 0.04 per cent.

   Sulfur, not over 0.05 per cent.
5. Tensile strength, pounds per square inch, 58000–68000.

   Yield point, minimum pounds per square inch, 0.5 tensile strength.

   Elongation in 8 in., minimum per cent., $\frac{1{,}500{,}000}{\text{tensile strength}}$.
6. Yield point by drop of beam method.
7. Cold-bend test; no cracking on outside of bent portion when bent 180° around a pin the diameter of which is equal to the thickness of the plate

5. *Cutting of Plates.*—(*a*) The method of cutting a plate 5 ft. by 20 ft. into pieces for welding is shown in the accompanying reproduction of drawing No. 102 of the Bureau of Standards.

(*b*) Each plate shall be given a distinguishing letter, A, B, etc. This letter shall appear upon each piece cut from it.

(*c*) Pieces shall be plainly marked with the letter and number arranged as shown in the drawing so that the location of the piece in the plate may be determined.

(*d*) The plates from which tensile, cold-bend, and fatigue specimens are to be made, shall be cut into pieces 9 in. by 30 in. (see Fig. 12).

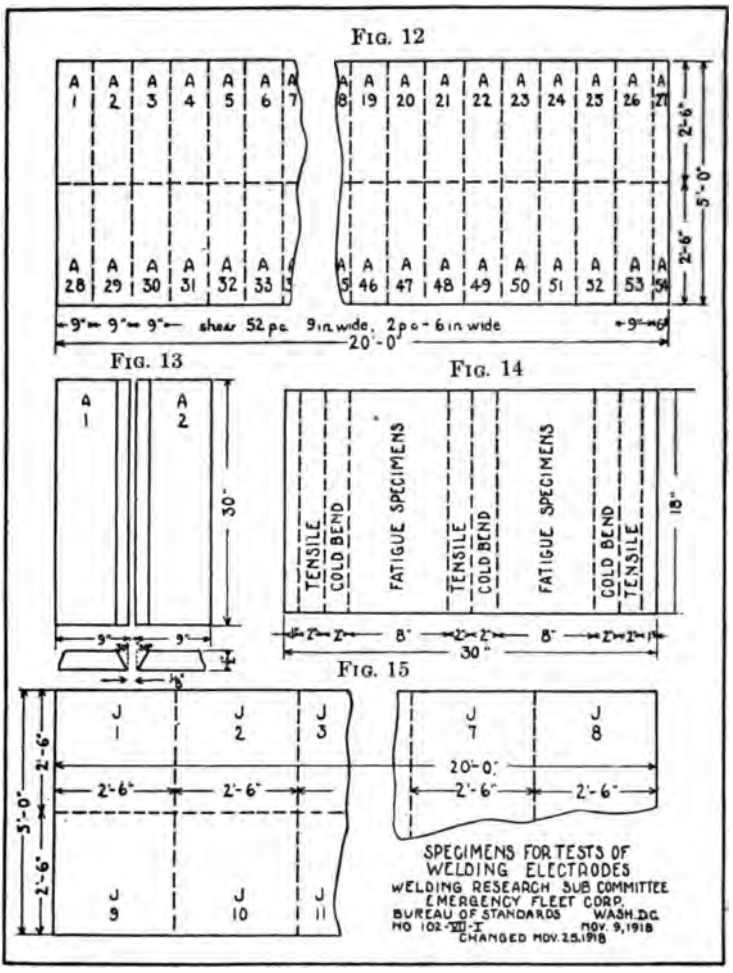
(*e*) The plates from which impact specimens are to be made shall be cut into pieces 30 in. by 30 in. (see Fig. 15).

Notes.—A power shear may be employed for cutting these plates if they are left flat. Pieces whose dimensions are within ¼ in. of the required size will be satisfactory.

6. *Arrangement of Pieces for Welding.*—(*a*) All welds are to be made across the grain, that is the welded joint is to be at right angles to the direction of rolling. Fig. 13 shows the correct arrangement of the pieces for the test welds to be used for the physical tests, but in Fig. 15, which shows the arrangement of pieces for the impact test, pieces such as J–1 and J–2 should be taken for a test weld instead of J–1 and J–9.

(*b*) Preparatory to making a weld, one 30-in. edge of each piece

will be beveled at the angle preferred by the applicant. If he has no preference, the angle of the bevel shall be 30° as shown in Fig. 13. The bevel is not to be carried through to the bottom edge but only to within $\frac{1}{16}$ in. of the bottom edge. In other words, the ends of the pieces to be welded are not to be beveled to a sharp point but $\frac{1}{16}$ in. of the original square edge will be left at the bottom of the vee.



(*c*) For welding, the pieces shall be placed horizontally with their under surfaces raised approximately $\frac{1}{8}$ in. above the supporting surface. The beveled edges shall be placed parallel and separated $\frac{1}{8}$ inch.

(*d*) The weld shall in all cases be of the type known as a double bevel.

7. *Welding of Pieces.*—(*a*) The pieces, which shall be dry and free from rust and foreign substance, shall be "tacked" at each end and in

the middle of the joint by welding the plates at each of these places for about ½ inch.

(*b*) After "tacking" the weld shall be completed by working continuously from one end of the joint to the other. The welding material shall be added in at least two layers; and when completed the surface of the weld shall be at least flush with and not more than ⅛ in. above the upper surface of the plate. The width of the weld at the top shall not be greater than 1¼ in. All welding shall be done from the open side of the vee.

(*c*) After each layer of a weld is completed, it shall be inspected by the authorized representative. If unsatisfactory, a new test weld shall be substituted and a report made of reasons for rejection.

8. *Number of Test Welds.*—(*a*) The welds will be made by three skilled operators, two of whom will be furnished by the applicant and one by the Welding Committee.

(*b*) Each operator shall make one 30-in. weld for the tensile, cold-bend and fatigue tests (see Fig. 13) using direct current and a similar weld using alternating current. For these tests, therefore, six test welds will be required for each sample electrode.

(*c*) Each operator shall also make one 30-in. test weld for the impact test (see Fig. 15) using direct current and a similar weld using alternating current. For impact tests, therefore, six test welds will be required for each sample electrode.

(*d*) If, however, the applicant intends his electrode for use with only one kind of current, that kind only shall be used in making the welds. In that case only three test welds will be required for (*b*) and for (*c*) instead of six.

9. *Arrangement of Conductors.*—(*a*) One of the conductors shall be attached to a clamp which will be fastened to both pieces at the end of the joint opposite that from which work is started.

(*b*) For direct-current work, the electrode shall have the polarity desired by the applicant. In the absence of any preference on his part, the instructions of the authorized representative shall be followed.

10. *Welding Data.*—The following information and observations shall be recorded by the authorized representative during the welding operations:

(*a*) Trade name or other identification mark of the electrodes.

(*b*) Complete description of the electrode.

(*c*) Sufficient description to identify the welding apparatus or system employed.

(*d*) Identification marks on the pieces consumed being welded.

(*e*) Name of operator.

(*f*) Kind of current (*i.e.*, direct or alternating); if direct, the polarity shall be given and if alternating, the frequency.

(*g*) Electrical quantities as follows: arc volts (both open and closed circuit), arc current and arc watts.

(*h*) Room temperature and humidity.

(*i*) The opinion of the authorized representative of the working quality of the electrode. This statement to include a description of any sputtering, boiling or other noticeable peculiarities.

(*j*) Elapsed time per weld.

(*k*) Weight of electrode consumed per weld.

(*l*) Any other information which will assist in determining the performance of the electrode, such as a photograph of at least one weld, etc.

11. *Preparation of Welded Plates and of Specimens for Physical Tests* (see Fig. 14).—(*a*) Each test weld shall be machined down on both sides to about the surface of the plate.

(*b*) Specimens shall be cut from each test weld reserved for physical tests as follows:

Three Tensile Specimens.—These shall be machined to a uniform width of 1.5 in. unless a weld of great strength makes it necessary to leave shoulders at the ends, in which case the standard A. S. T. M. test specimens for sheet iron and steel shall be prepared.

Three Cold-bend Specimens.—These shall be machined to a uniform width of 1.5 inches.

Six Fatigue Specimens.—These shall be machined to about ½ in. diameter and 10 in. long. (The exact dimensions are to be determined by experiment.)

12. *Physical Tests.* (*a*) *Tensile Strength.*—Each of the three specimens shall be tested in accordance with the practice recommended by the A. S. T. M. and shall include the determination of the tensile strength, yield point (by drop-of-beam method), reduction of area and total elongation after rupture in 2 and in 8 inches.

(*b*) *Cold-bend Test.*—This test shall be made by placing the specimen on two rollers with the apex of the vee upward and midway between the rollers and loaded at the center of the span thus formed by a cylindrical surface having a diameter of ½ in. This surface shall bend the specimen downward between the rollers until a fracture appears on the lower side of the specimen when loading shall be stopped and the angle noted through which the specimen has been bent.

(*c*) *Fatigue Test.*—Each of the six specimens shall be tested in a special rotating type of machine similar to that used by Lloyd's Register of shipping. (Exact details to be determined by experiment.)

(*d*) *Impact Test.*—Each impact-test specimen shall be placed on supports 18 in. high and 4½ ft. apart. A spherical weight of 500 lb. shall be allowed to fall freely through a distance of 10 ft. striking the weld which shall be at the center of the span. The apex of the vee shall be upward.

(*e*) *Test of Original Plate.*—In order to establish the physical properties of the unwelded plate, tensile, cold-bend and fatigue tests shall be made on a sample selected at random from the pieces used for the test welds but before such welds are made.

Optional Tests

The following tests are optional with the applicant but the Welding Committee considers that they give information of great importance and recommends that they be made in order to make the report on the electrode more complete.

13. *Chemical Analysis.*—A chemical analysis shall be made of: (*a*) The original plate in one test weld selected at random; (*b*) the metal at the center of one test weld selected at random. If test welds are made with both direct and alternating current, an analysis shall be made of one weld of each set.

14. *Photomicrographs.* — Photomicrographs shall be made of one specimen weld selected at random as follows: (If test welds are made with both direct and alternating current, photomicrographs shall be made of one weld of each set.) (*a*) At center of weld, (*b*) at juncture of weld and original metal, (*c*) in adjacent original metal, (*d*) cross-section of electrode, (*e*) longitudinal section of electrode.

APPENDIX B

Wirt-Jones Tests of Arc-welded ½-in. Ship Plate

In Tables 4 and 5 are set forth the results of the tests carried out in 1918 on arc-welded ½-in. ship plate by the Welding Research Subcommittee. The plates with which the welds were made had their edges prepared with a 45° double vee. The welds were made by a dozen different concerns. In each instance the operator employed whatever conditions were considered most suitable. Several types and sizes of electrodes, including both bare and covered, were used. While in the majority of cases direct current was employed, alternating currents of 25 and 60 cycles are also represented in the series. The number of layers and the current employed and, indeed, all the conditions, were left to the discretion of the individual operator. The length of the weld was 8 in. After the welds were made, the plates were cut up by the Wirt Company (through the courtesy of Mr. Charles Wirt). The tests were made at the Bureau of Standards under the direction of Prof. H. L. Whittemore. The tensile tests were made on specimens with

a cross-section of 1 in. by ½ in. Specimens with the weld machined have the projecting metal planed off so that the welded portion is smooth and of approximately the same cross-section as the remainder of the specimen. For the specimens with the weld not machined, no record was made of the added cross-section due to the projecting material. The cold-bend specimens were 1 in. by ½ in. and were bent around a mandrel of 1.5 in. diameter. The results of the tensile and bending tests for the machined samples are given in Table 4. Those for the samples which were not machined are given in Table 5.

TABLE 4.—*Wirt-Jones Tests of ½-in. Arc-welded Ship Plates —Machined Samples*

Test Number	Current, in Amperes		Diam. of Elec. in In.	Current Density, Amp. per Sq. In.	Tensile Tests		Bending		Electrode	
	A. C.	D. C.			Ultimate Strength, Lb. per Sq. In.	Per Cent Elon. in 2 In.	Angle at which Crack Starts. Degrees	Cov.	Bare	Make
20	175	...	0.125	14,227	62,700	9.0	34	C	..	E. A. C. & W. Co.
1	...	175	0.166	9,118	59,800	9.0	42	..	B	Roebling
22	160	...	0.125	13,008	50,000	6.0	78	C	..	E. A. C. & W. Co.
2	...	155	0.156	8,075	62,600	11.5	44	..	B	Roebling
19	150	...	0.125	12,195	62,800	12.0	32	C	..	Quasi
31	...	150	0.166	6,930	60,900	8.0	..	..	B	Roebling
29	...	150	0.156	7,815	56,400	6.5	..	..	B	W. W. & M Co.
38	150	...	0.156	7,815	54,500	7.0	45	C	..	E. A. C. & W. Co.
29	...	150	0.156	7,815	54,400	6.0	47	..	B	W. W. & M. Co.
9	...	150	0.156	7,815	44,200	4.0	26	..	B	Toncan
9A	150	...	0.187	5,430	42,500	4.0	30	..	B	Armco
10	...	150	0.156	7,815	41,900	4.5	25	..	B	Toncan
10	...	150	0.156	7,815	41,300	4.5	24	..	B	Toncan
9	...	150	0.156	7,815	38,100	4.0	15	..	B	Toncan
9A	150	...	0.187	5,430	36,300	4.0	..	..	B	Armco
5	...	145	0.156	7,554	61,100	8.5	55	..	B	Am. S. & W. Co.
25	136	...	0.156	7,085	43,700	3.5	26	..	B	Roebling
30	...	135	0.156	7,033	53,200	11.0	..	..	B	W. W. & M. Co.
30	...	135	0.156	7,033	52,000	10.5	50	..	B	W. W. & M. Co.
37	125	...	0.125	10,162	56,400	7.0	42	C	..	E. A. C. & W. Co.
7	...	120	0.125	9,756	57,600	8.5	50	..	B	Roebling
7	...	120	0.125	9,756	53,700	7.0	45	..	B	Roebling
8	...	115	0.125	9,349	59,400	13.5	60	..	B	Roebling
8	...	115	0.125	9,349	58,200	14.0	..	..	B	Roebling
3	...	115	0.156	5,891	50,600	5.0	34	..	B	Am. S. & W. Co.
16	...	110	0.134	7,799	50,250	7.0	35	C	..	Quasi
27	...	110	0.125	8,943	45,800	8.0	42	..	B	Roebling
21	105	...	0.134	7,444	42,200	5.0	42	C	..	E. A. C. & W. Co.
26	94	...	0.125	7,642	34,200	3.0	15	..	B	Roebling
18	90	...	0.134	6,381	54,000	6.0	65	C	..	Quasi
15	...	90	0.134	6,381	39,400	4.0	30	C	..	Quasi
24	...	86	0.125	7,154	45,400	6.0	15	..	B	Roebling
23	...	85	0.125	6,930	51,900	4.0	..	..	B	Roebling

TABLE 5.—*Wirt-Jones Tests of ½-in. Arc-welded Ship Plates —Samples Not Machined*

Test Number	Current, in Amperes		Diam. of Elec in Inches	Tensile Tests			Electrode		
	A. C.	D. C.		Current Density, Amp. per Sq. In.	Ultimate Strength, Lb. per Sq. In.	Per Cent. Elon. in 2 In.	Cov.	Bare	Make of Electrode
20	175	...	0.125	14,275	66,480	11.0	C	..	E. A. C. & W. Co.
22	160	...	0.125	13,008	66,400	9.0	C	..	E. A. C. & W. Co.
19	150	...	0.125	12,195	65,400	6.0	C	..	Quasi
5	...	145	0.156	7,554	52,280	7.0	..	B	Am. S. & W. Co
25	136	...	0.156	7,085	47,800	4.0	..	B	Roebling
3	...	115	0.156	5,891	65,470	13.0	..	B	Am. S. & W. Co
27	...	110	0.125	8,943	48,200	7.0	..	B	Roebling
16	...	90	0.134	6,381	58,740	5.2	C	..	Quasi
18	90	...	0.134	6,381	61,760	11.0	C	..	Quasi
15	...	90	0.134	6,381	57,340	8.0	C	..	Quasi
23	...	85	0.125	6,910	57,060	5.0	..	B	Roebling

APPENDIX C

BENDING TESTS OF GAS WELDS

Too late for inclusion in the body of this paper, there has been brought to the author's attention some gas welds made under the direction of

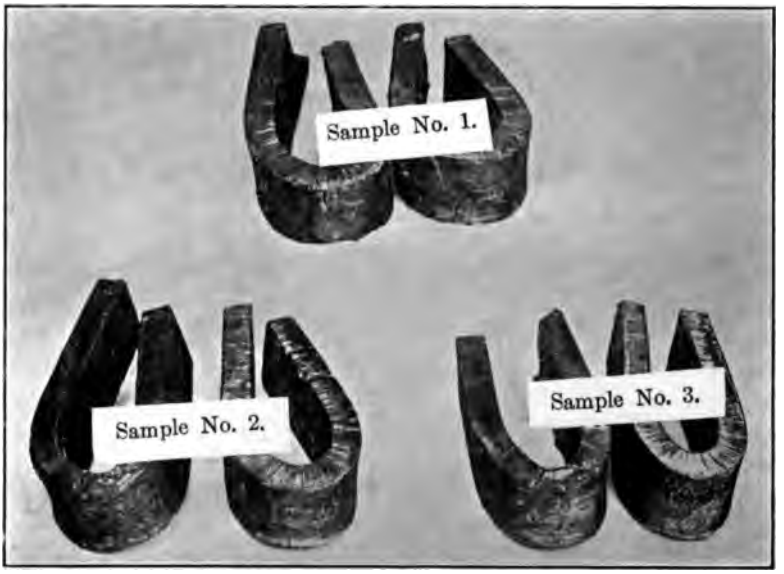


FIG. 16.

H. I. Walsh of the Newport News Shipbuilding Co., which have withstood very severe bending tests before failing. The specimens are shown

in Fig. 16. Three pairs of specimens welded with three kinds of electrodes are shown. These specimens show much greater ductility than those shown in Fig. 6.

APPENDIX D

Arc-welded Steel Box

Mr. R. E. Wagner has prepared the following description of a large steel box, arc welded with bare electrodes at the Pittsfield Works of the General Electric Co. The box was built and tested in the latter half of 1918. It was made of ½-in. tank steel and its dimensions were 12 ft. by 10 ft. by 9 ft. The total weight was 16,000 lb. The welding was

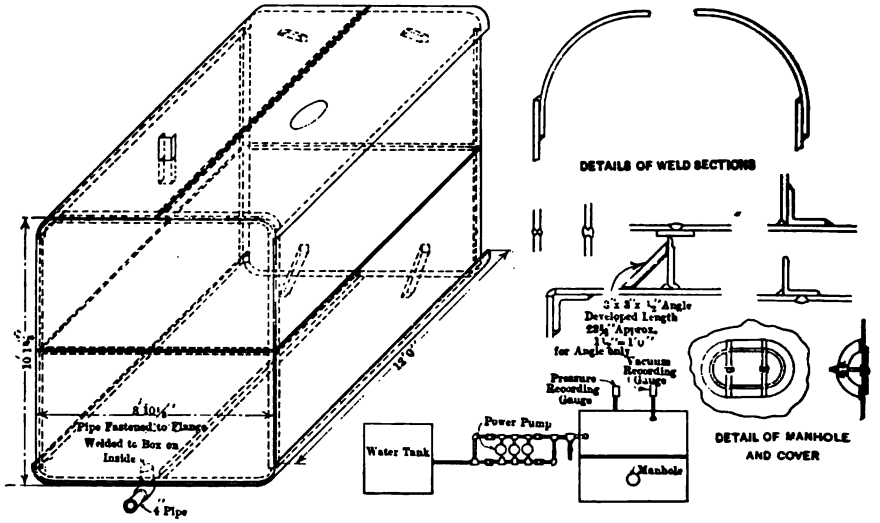


Fig. 17.—Outline and dimensions of arc-welded steel box.

done with 75-volt direct current. This box contained eleven varieties of welds. Its general dimensions and the types of welds used are shown in Fig. 17. No rivets were used in its construction.

The principal information sought in the construction of this box were:

1. Could a structure of this character be made close to drawing dimensions and without excessive distortions and warping of plates and parts?

2. Would the structure be strong and capable of withstanding severe shocks and distortions without serious ruptures?

3. What are the detail costs and time required to build such a structure?

It was found that if proper allowances are made for contraction and expansion, especially where plates are butted and welded, the resulting

FIG. 19.—ELECTRIC-ARC WELDED BOX, UNDER 22-IN. VACUUM, SHOWING TOP SIDE DEPRESSED.

FIG. 18.—ELECTRIC-ARC WELDED STEEL BOX OF ½-IN. TANK STEEL.

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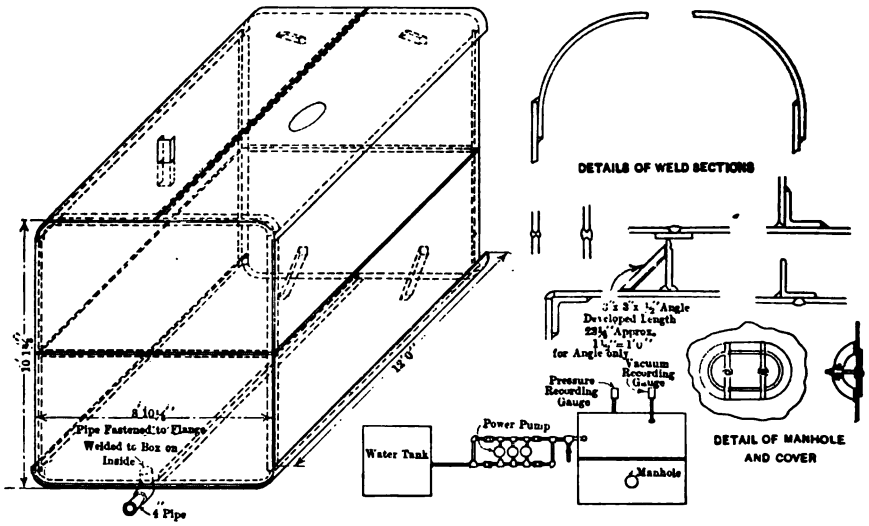

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FIG. 19.—ELECTRIC-ARC WELDED BOX, UNDER 22-IN. VACUUM, SHOWING TOP SIDE DEPRESSED.

FIG. 18.—ELECTRIC-ARC WELDED STEEL BOX OF ¼-IN. TANK STEEL.

prevents much loss of carbon. The larger globules tend to increase the arc fluctuations due to greater partial short-circuiting of the arc. Actual oscillographic records show that with the hot-rolled steel electrodes the current peaks were 70 per cent. above normal whereas the peaks for the other electrodes were only about 25 per cent. above normal. Probably due to the larger globules, a greater proportion of slag in the electrode was carried into the weld. Some four or five test pieces made with the hot-rolled steel electrodes were all considerably below the tensile strength of the samples welded with the other electrodes. These facts indicate that an electrode may not be satisfactory even though of approximately the proper chemical composition. For example, the hot-rolled steel and Roebling electrodes are quite similar in chemical analysis.

TABLE 6.—*Speed Obtained in Arc Welding Ship Plates*

Test No.	Speed, in Feet per Hour	Current Used	
		Amperes	Cycle
1	3.33	170–180	
2	2.86	150–160	
19	1.90	150	60
20	1.48	175	60
21	1.59	105	25
22	0.97	160	25

On page 583, the relative speeds of alternating- and direct-current arc welding are mentioned. By referring to Table 4 on page 608 we find that tests 1 and 2 were made with direct current, that tests 19, 20, 21 and 22 were made with alternating current. The speeds of welding are shown in Table 6. Of course these are isolated cases, but it is interesting to see what is obtained from actual tests.

On page 593, Mr. Hobart has brought out the question of suitable current values. This is probably one of the most important and least understood problems in the entire arc-welding category. In fact, it is the problem of proper fusion in arc welding, which can only be secured by the use of the proper current volumes, depending on the characteristics and temperature of the material to be welded, and by the use of the proper diameter of electrodes for the current. A visual inspection of the deposited metal will almost unfailingly indicate when the desired effects are obtained. The method of determination is very ably and clearly pointed out in Mr. Eschholz's[1] paper; namely, judging the penetration metal by the depth of the arc craters, measured from a straight line

[1] Presented at session with American Institute of Electrical Engineers. Printed in A. I. E. E. *Proceedings* (1919) **38**, 319.

across the crater top to the bottom of the crater, and an examination of the edge of the deposits for re-entrant angles, to see how much of an angle there is.

Mr. Hobart also made reference to the effect on arc welding of voltage employed. I do not believe that the question of voltage is quite as simple as inferred, unless coated or wrapped metal electrodes or carbon electrodes are used. We will grant that for most open work an experienced conscientious welder can produce results with circuit potentials as high as 250 volts; however, an inexperienced or a careless operator will follow the course of least effort and maintain a long arc, resulting in an unconcentrated porous deposit, which will be poorly fused and does not penetrate into the shank metal. An inexperienced operator certainly ought not to be trusted with such high open-circuit voltages, as he is sure to produce unsatisfactory welds.

C. J. HOLSLAG,* Newark, N. J.—Speed in welding may be advantageous or disadvantageous according as to whether the metal is deposited faster than the operator can make sure it is being deposited in the crater made fluid by the arc, or whether some of the metal is getting ahead or outside of the crater. Speed at which good welding can be done is an advantage and alternating current provides a greater speed than direct-current welding with the electrode negative, because with the electrode negative approximately 60 per cent. of the heat is in the work, and with alternating current the heat is naturally fifty-fifty between the electrode and work.

Direct current with the electrode positive provides faster welding, but generally too fast to be of any use, as mentioned, unless the arc is closed in by a very heavy coating so that the metal is enclosed in a viscous sleeve, or where the work is very rough, such as filling in with a large electrode on a casting where strength is not required. In a test recently made with direct-current and alternating-current machines, the rate of melting was taken; that is, the number of pounds per hour that could be melted with no stops was determined. In each case under the same conditions, *i.e.*, different currents and voltages were taken, but comparisons made only when they were the same, alternating-current melting was from 20 to 30 per cent. faster. For instance, at 150 amp. and 25 volts, 4 lb. an hr. were deposited with alternating current and 3 lb. an hr. with direct current; and at 175 amp. and 25 volts, 5 lb. an hr. with alternating current and 4 lb. an hr. with direct current. With a $\frac{5}{32}$-in. electrode, from 4 to 5 lb. an hr. is the maximum amount that can be deposited in electric arc welding, for with higher current densities good welding is lost.

Speed also depends on the length of the arc because the rate of melting depends on the rate of heating and the rate of heating naturally de-

* Electric Arc Cutting & Welding Co.

pends on the voltage across the arc, as well as the current through it. With a long arc, this rate can be raised, and hence the welding speeded up; but as pointed out in Mr. Eschholz's paper, everything is in favor of a short arc for good welding, *i.e.*, less air radiation, less contact with the air for oxidation and forming of nitrides, greater chance that metal of the electrode will deposit in the crater and nowhere else. With alternating-current apparatus, the greater possibilities for automatically holding a short arc, and constant rate of heating are generally taken advantage of. Hence any test of speed must be accompanied by voltage across the arc and amperes through it, which means by definition, the same rate of heating; and with the same rate of heating, the only factor left is the disposition of the heat at the positive and negative electrodes, for as explained before, the amount of heat at the electrode end determines the rate of its melting.

W. L. MERRILL,* Schenectady, N. Y.—Until about 2 years ago, spot welding was limited to very thin sections, say up to ¼ or ⅜ in. (6 or 9 mm.) thick; beyond that the field had not been thoroughly explored. It was thought that it might be advantageous to use this method in the construction of ships, so tests were made and a machine built; the results have been published in Major Caldwell's report and in one issue of the *General Electric Review*, which has been referred to by Mr. Hobart. As a result of those tests, it was found possible economically to build machinery capable of welding two or more pieces of structural steel up to a total of 3 in., that being the maximum thickness we tried to weld.

Unlike arc welding, spot welding has very few variables; and a surface inspection in practically all cases is all that is necessary to determine a 100-per cent. weld. Also, like the arc weld that we have been hearing about today, spot welding can be done to more than 100 per cent. strength of the plate. As an example, page 597 contains some test samples that were made up of different kinds of welding. The two-spot weldings, which would be used for these thicknesses of plates, gave a breaking strength of 28,000 lb. as against 13,000 lb. for the standard riveting for that same plate.

As a result of experiments carried on in the large machine, which by the way was 2000-kv-a. capacity, although we did not reach the capacity of the machine, three separate and distinct machines were designed for use in fabricating ships; that is work to be done in the shop, not on the ships. Pictures and explanations of these are given in Mr. Hobart's paper. These machines have been completed, and two of them are in the possession of the Emergency Fleet Corpn. at the present time, working out the practicability of adapting them to the fabrication of ship parts in place of laying out, drilling, and riveting. Only experience will tell

*Engineer, Power & Mining Engng Dept., General Electric Co.

what saving or gain, if any, there will be in the use of this type of machinery. One thing, however, we know can be gained, and that is for a given strength of a given member, the joints will not be limiting features, and if 60 per cent. is figured for riveted joints, the thickness of the plate can be reduced proportionately because you are always sure of 100 or more per cent. strength if spot welding is done with proper care.

D. C. ALEXANDER, JR.,* Brooklyn, N. Y.—In mining work, manganese steel is used for rails, crossovers, and, I believe in some cases, wheels on mining cars. Now, for that sort of work it has been customary, when the manganese steel parts wore out, to scrap them because there seemed to be no way of repairing them. A covered electrode, if it is properly covered, will deposit the slag on the molten metal so that the constituents of the original metal can be retained in the positive metal. So by starting with a high manganese-steel wire and putting the suitable covering on that wire, it is possible to deposit on a plate, or manganese specimen of some sort, 12 per cent. manganese steel. That in itself justifies the covered electrode for that particular kind of work. I imagine mining engineers may have occasion to build up worn parts composed of high-carbon steel. The same effect can be produced by using a 0.7 per cent. carbon steel wire, and depositing 0.5 per cent. carbon steel, to build up carbon steel parts.

The remarks made this afternoon about covered electrodes have been a little bit vague in some cases, and possibly some of you have the idea that any kind of covering on an electrode produces the same result. I noticed in Mr. Hobart's paper that, in some instances where it did not give satisfactory results, a great improvement could be made by dipping the wire into the melted iron. That may be correct, and some people may think they will get a big improvement in that way. But when you think about it, you must realize that in steel-furnace practice you must have a suitable slide, you cannot have any kind of slide; similarly the material used as electrodes must not be selected because it is cheap or common, but because it is the only one found to contain the proper constituents. What we use is blue asbestos, which is not like white asbestos at all. The coating fuses with the steel and covers the molten steel with a thin slag, which keeps out the oxygen and prevents the formation of oxide. It also maintains in the molten metal the constituents of the steel that we want to keep there.

W. SPRAGUE,† New York, N. Y.—The angle referred to in Table 3 is the angle on each plate, therefore the total opening in each weld will be just double that given. For instance, the total opening for the 15° bevel listed in this table would be 30°. The statement is in con-

* Quasi-Arc Weldtrode Co., Inc.

† Electric Welding Research Assistant, American Welding Society.

nection with article 24, on page 594, and Mr. Hobart states that no physical test has been made on these samples. I have, however, received from the Electrical Testing Laboratories some results of tensile and bending tests on these specimens; each result given below is an average of three test specimens.

Angle of Bevel on Plates, in Degrees	Total Opening, in Degrees	Ultimate Strength, in Lb. per Sq. In.	Angle of Bevel at which Crack Starts, in Degrees
15	30	44,000	9
30	60	51,000	15
45	90	45,300	13
60	120	48,800	14

These results seem to indicate that the advantages claimed by the American traditions for the very large opening are rather questionable if we also take into account the time, electrode material, and power saved by the smaller angle of bevel, as shown in Table 3. Obviously one could have too small an opening so that the operator would not have sufficient room in which to properly manipulate his electrode. The results of the above test would indicate that this point was reached when the total opening was 30°.

J. C. ARMOR, Pittsfield, Mass. (written discussion*).—Mr. Hobart has presented a comprehensive list of points that are of vital interest in fusion welding. In alternating-current arc welding of steel plates with bare electrodes, good sound welds can be produced with wire that is composed of practically pure iron alloyed with from 0.02 to 0.65 per cent. manganese. The brands of welding wire that contain 0.30 to 0.60 per cent. manganese will generally produce the strongest welds, while those carrying a very small amount of manganese will, in general, produce the softest welds. I have not had much success with wires carrying more than 0.65 per cent. manganese or with alloying elements other than manganese.

There is very little, if any, difference in the quality of work done with alternating current and direct current, provided, of course, that proper conditions are maintained in each case. It is as easy to work with alternating current as with direct current, provided the alternating-current "striking" voltage is sufficiently high. Any commercial frequency is suitable for arc welding if sufficient open-circuit voltage is employed. The lower frequencies require somewhat higher voltages. My experience indicates that the speed of welding is about the same with alternating current and direct current. An operator who is ex-

* Received Feb. 19, 1919.

perienced in direct-current welding usually adjusts his alternating current to a value that allows him to manipulate the fused metal in the way he has always done with direct current, with the result that he melts in about the same quantity in a given time.

In connection with Mr. Hobart's 20th point, Effect on Arc Welding of Voltage Employed, it is, of course, a well-known fact that the alternating-current arc is much more easily extinguished than the direct-current arc and also that the lower the frequency the more easily is the arc interrupted. These facts must be taken into account in welding with alternating current. The general tendency has been to employ alternating potentials of too low a value for effective work, with the result that the operator cannot manipulate his arc properly and therefore has difficulty in producing good work. This question of open-circuit potential is extremely important; and if it is desired to be free to use all welding wires now on the market, the "striking" voltage for 60 cycles should be at least 150 volts. Good welds can be made with less voltage when using certain brands of wire but other brands will require 150 volts. When the voltage is too low, there is a very strong tendency to hold the arc between the electrode and the top, or outside, of the weld, with the result that the metal at the bottom of the weld is not fused in as well as it should be. Arc welding, even under the best possible conditions, is very trying to a conscientious operator and for that reason the work should be done in the easiest practicable manner, as it is impossible for an operator to do uniformly good work if he is under a continuous strain. While an expert may make a few very good test welds, using a potential so low that he has the greatest difficulty in working, he will not be willing to work that way day after day; and if required to do so he will tend to become discontented and careless. For the reason just given, the alternating-current potential employed for most commercial wires now on the market should, in the writer's opinion, be about 140 or 150 volts.

In regard to Mr. Hobart's 28th point, Consequences of Different Lengths of Arc, the writer's experience has been that the shortest practicable arc produces the best welds. A long arc, on the other hand, deposits inferior and weak metal in the weld. No doubt this is due to the absorption of larger amounts of oxygen and nitrogen from the air. The question of arc length is certainly as important as the electrode composition and the secret of good welding with most of the commercial wires now on the market is found in the maintenance of a very short arc manipulated so that the deposited metal fuses into the plate over the whole surface to be welded.

S. V. GOODALL* (written discussion†).—The thanks of those interested in welding are due to Mr. Hobart for this valuable paper.

Mr. Hobart omits to refer to the investigations carried out by the

*Constructor-Commander, [redacted]. *Received May 5, 1919.

Welding Research Sub-committee to ascertain whether any trouble is to be expected from excessive corrosion of a welded joint exposed to sea water. The point was raised, early in 1918, that on account of the difference in chemical composition and electrolytic characteristics of mild-steel ship plate and the metal laid on at the weld, it is possible that excessive corrosion would set in at the joint if the protective and anti-fouling compositions were accidentally removed or otherwise rendered ineffective, and the joint might fail in consequence. As welding had already been employed for effecting repairs to under-water parts of ships, the Welding Research Sub-committee instituted inquiries to ascertain whether excessive corrosion had been experienced in these cases; and while few replies were received, owing to the fact that the history of the weld after the repair was effected had not been kept, no actual cases were cited where excessive corrosion had occurred.

The British Admiralty and the U. S. Navy Department are carefully watching the effect of sea water on welded joints, and under the superintendence of Commander H. G. Knox of the U. S. Construction Corps, a series of tests on the corrosion of welded joints made with various types of electrodes is being carried out at the Norfolk Navy Yard. The Bureau of Standards is also investigating this matter. So far the experience at present available and the opinions obtained lead to the conclusion that no danger need be feared from excessive corrosion of a weld on service; further experience, however, is considered necessary before this question can be regarded as definitely settled.

With regard to the application of spot welding to shipbuilding, Mr. Hobart states that it is planned that bulkheads, frames, floors, and other parts shall be spot-welded and then transported to their places in the ship. This sounds attractive; but the shipbuilder will at once ask how the cost of such spot welding compares with that of riveting. The riveting of parts such as those mentioned can be done in shops or sheds, and is comparatively inexpensive. If it is to be displaced by spot welding, this latter process must be considerably cheaper than I understand it to be; and if Mr. Hobart could add particulars as to the relative cost of spot welding and riveting such members, it would add to the value of this part of the paper. With regard to the adoption of spot welding for work at the ship, there are many difficulties ahead for those shipbuilders who attempt this work; this opinion is based on experience with portable hydraulic riveters in ship construction and is altogether apart from the practicability, or otherwise, of efficiently connecting two heavy plates by spot welding. Even in this latter respect it must be remembered that there is considerable difference between the spot welding of two small thick plane plates in a shop and the spot welding of two large non-planar plates forming the complex structure of a ship. Moreover, spot welding, compared with line welding, possesses the disadvantage that economy of steel is not likely to be so great.

Mr. Hobart, in presenting the paper, rather ineffectively trailed the coat-tails of the bare-wire advocates in the face of those who believe in covered electrodes. In Great Britain, shipbuilders and engineers are not paying the additional price of covered electrodes for philanthropic reasons; as soon as they are convinced that the bare electrode will give as good and as uniform results as the more costly covered wire, Mr. Hobart may rest assured that the bare electrode will be freely used in that country. But I am bound to say that present experience in Great Britain indicates that the covered electrode gives superior results and gives such results more uniformly, the superiority being particularly indicated in fatigue tests, which is an extremely important matter if welding is to be employed on parts of ships contributing to general structural strength. Mr. Hobart states that one type of electrode must yield results uniformly superior to those obtained with another type in order to afford economic justification for a five times greater price. It is not considered that this matter can be summed up in this manner in regard to shipbuilding if electric welding is employed on such parts that the resulting joint is of importance in the general structural strength. If, by using a high-priced electrode, the desired minimum standard of strength can be always attained with certainty, while the lower priced electrode cannot be relied on to attain such a standard at all times, the shipbuilder would not be justified in using the cheaper article, unless it were possible to subject all the joints to some test that would imitate in quantity and quality the forces that will be brought upon the joints on service. The cost of such tests might offset the lower cost of bare electrodes. Moreover, if two different types of electrodes were used on a ship—the cheaper where strength is not of much importance and the more expensive where structural strength is vital—extremely careful supervision would be necessary, and it would probably be more practicable for the shipbuilder to adopt the higher priced article throughout.

It is desired to emphasize point 30 of Mr. Hobart's paper, on the necessity of a clean surface for fusion welding. Some of the unsatisfactory work that I have seen is due to the attempt to weld together plates, etc., having surfaces that are oily or are otherwise insufficiently clean.

R. P. JACKSON,* Pittsburgh, Pa. (written discussion†).—Both the varied and somewhat conflicting information set forth in Mr. Hobart's paper and that obtained by observation and experience indicate that in a large measure electric welding is an art rather than a science. While it is obvious that a great deal can be done in the way of eliminating the useless and fake elements, after all that a scientific investigation can contribute has been taken into account, the actual production of safe welds is essentially a personal problem. A skillful welder will do some very

*Research Division, Westinghouse Elec. & Mfg. Co. †Received Mar. 6, 1919.

creditable work with a variety of materials and equipment and an unskillful one will do poor and unreliable work with the most perfect equipment.

It is this feature of the essential dependence on the skill of the individual welder that has been such a handicap to the more extended use of welding. The difficulty is that a designing engineer laying out his computations and drawings cannot know with any degree of certainty what he can depend on in the way of a joint as compared with riveted work put up with a very moderate degree of skill but capable of being checked by inspection.

While this comment does not disparage the necessity of research work and particularly of a greater degree of publicity as to the actual facts, its purpose is to accentuate the necessity of training welders and inspectors. The writer feels that no extensive undertaking like shipbuilding can be worked out except as a result of the building up of a welding organization, comprising a superintendent, inspectors and welders who are conscientious and know their job. Without such an organization, no cautious engineer would specify the use of welding for the obvious reason that he could not be sure what he would get. With such an organization, on the other hand, there is almost no engineering structure that could not be undertaken, with the only limitation as to the amount of welding to be the accessibility of the assembled structure for the proper welding operation.

F. M. FARMER,* New York (written discussion†).—The paper presents many phases of the welding art concerning which there are very divergent views. In many cases, the wide differences of opinion are based on the results of mechanical tests, but an examination of the mechanical test data discloses a wide variation in character of the tests made and in the details of procedure, so that it is frequently not only difficult to analyze and check the basis of the conclusions but impossible to correlate various groups of test results. It would seem, therefore, that the technique of testing welds is one of the phases of the welding art that should be studied and standardized in order that the results of mechanical tests made by different people may be strictly comparable and have the same significance to all.

The vital difference between testing a specimen of steel that includes a welded joint and testing an ordinary specimen is the non-homogeneity of the weld specimen. The weld specimen has at its center a section composed of material that usually has physical, chemical, and metallurgical characteristics distinctly different from the adjoining metal. Furthermore, the section of added metal is more or less irregular and the ratio of added metal to original metal in the section is quite variable. Consequently the details of procedure and of recording data should be more minutely prescribed than in the testing of ordinary specimens.

* Chief engineer, Electrical Testing Laboratories. † Received Mar. 10, 1919.

A conspicuous example of a mechanical test that requires standardization is the cold bending test, a standard test for mild steel. With the usual method, the specimen is bent around a pin of prescribed diameter and, as all specifications usually require bending to 180° without fracture, the manner of applying the load is immaterial because the specimen is ultimately completely wrapped around the pin. But a specimen containing a welded joint at the center will usually not withstand bending to 180° because of the much lower ductility of the deposited metal in the weld. It is therefore obvious that the farther apart the points of support are, the greater will be the angle attained before cracking occurs, simply because most of the bending will take place in the original metal. It is therefore highly desirable that there be prescribed a standard test that fixes exactly the distance between the supports, the curvature of the surface of supports, and the relation between this curvature, that of the loading surface, and the thickness of the specimen.

The Research Sub-committee of the Welding Committee of the National Research Council is using, in connection with some of its research work, a bending test that tests a specimen 1½ in. wide as a simple beam with two supports and with the load applied at the center of the weld and midway between the supports. The load is applied to a cylindrical surface, about which the specimen is bent, that has a diameter equal to the thickness of the specimen, which corresponds to the specifications of the American Society for Testing Materials for cold bending tests of steel plate. The supporting surfaces are placed as close together as practicable so that the deposited metal in the weld will occupy the largest possible proportion of the beam. This minimum distance is obviously that which will permit the bending of a specimen to 180°. The corners of the supports are machined and polished to a radius equal to the diameter of the loading surface and to the thickness of the specimen. The initial distance between the points of support is then the sum of twice the radius of the corners of the supports, the diameter of the loading surface, twice the thickness of the specimen and ⅛ in., which gives 1/16 in. clearance on each side. Such a test is very easily made in a standard testing machine by attaching a suitably prepared die to the moving head of the machine and setting two blocks with rounded corners on the bed of the machine with the proper space between them. The first crack on the under side of the specimen is readily detected with the aid of a mirror.

In a considerable number of tests of weld specimens made by this method, the maximum angle obtained was 15° while the angles usually reported in the literature are from 30° to 60°. It seems certain that the relatively small angle found is due to the greater severity of the test and not to any inferiority of the welds. It should be stated that

the parent metal can be bent to 180° by this method without a sign of a crack.

Similar standardization is required in tensile tests, particularly the gage length. With long gage lengths, the elongation in the weld itself is obscured by the stretch in the parent metal.

Mr. Hobart referred to the feeling that has been frequently expressed that more information in regard to the durability of the welded joints under the action of repeated loading is needed before entire confidence in welding in certain fields can be obtained. A special fatigue testing machine was developed in connection with the work of the Research Subcommittee of the Welding Committee for testing specimens of welds in ½-in. plate. The machine employs the rotating-beam principle, the beam having two free supports at the ends and two loads at the middle. The arrangement is such that the stress is absolutely uniform throughout the length of the specimen occupied by the weld and the exact maximum fiber stress can be calculated. The principle is not at all new but the particular form of apparatus developed is very small, simple, and inexpensive. The results of tests recently made indicate that where standard ship-plate steel will withstand 4,000,000 or 5,000,000 reversals of stress at 25,000 lb. per sq. in. before failure, electrically welded joints will withstand 700,000 or 800,000 repetitions of the stress. All the failures occur in the deposited metal, indicating that there is no difficulty in getting a thorough bond between the added metal and the original metal, but that the added metal has not the fatigue strength of the original metal. This simply confirms the general view that greater ductility in the deposited metal of electric welds must be had before the process can be applied to a maximum extent in those very large fields where this characteristic is absolutely essential.

W. S. ANDREWS,* Schenectady, N. Y. (written discussion†).—Mr. Hobart has omitted to make any mention of colored glasses to be used by welders, in connection with goggles, shields, and masks for protecting their eyes and skin from the dangerous radiations produced by the electric arc. The iron arc is especially prolific in generating ultraviolet radiation, which will induce painful and dangerous inflammation in the unprotected eye, and a powerful arc will produce a sunburn effect on the bare skin in a very short time. Leather gloves on the hands and shields or masks of light material, such as thin wood or vulcanized fiber, afford a perfect protection to the face and neck from these harmful rays. The protection of the eyes from these rays and the intense glare of the arc, however, is not so easy, as it involves the necessity of preserving clear vision for the operator.

It is well known that the human eye is more sensitive to yellow

* Consulting Engineering Department, General Electric Co.

† Received Feb. 15, 1919.

and green light than it is to red and blue light; so, in order to secure the clearest vision with the least amount of glare, densely colored glasses should be used. These should exclude most of the red rays and practically all blue and violet rays, while the orange, yellow, and green rays should be toned down to the lowest degree consistent with clear observation of the work in hand. Owing to individual difference in visual acuteness, no absolute rule can be given as to the exact amount of obscuration necessary for various operations. This may be decided best by each operator for himself, for one man may be able to see clearly through a glass that would obscure the vision of another. On the other hand, the fundamental color transmission of all safety glasses should be determined by an expert in accordance with spectroscopic analysis. There are many safety glasses on the market that fully conform with these requirements. Glasses dense enough to tone down the intense glare of the electric arc to an extent comfortable to the eye will always afford sure protection against ultraviolet radiation.

Infra-red radiation or heat rays naturally produce a drying effect on the skin; but as the latter is sensitive to them, they give their own danger signal. The shields and masks worn as a protection against the ultraviolet rays will usually be found a sufficient guard against heat rays. Most of the reliable dense-colored safety glasses on the market are also good shields against heat rays, so there is very little danger, if any, to be apprehended from this source. All harmful effects to the eye from the radiations of the iron arc may indeed be entirely avoided by the exercise of a little common sense and intelligent caution.

J. C. LINCOLN,* Cleveland, Ohio (written discussion†).—The general problem of electric welding, as it presents itself to the writer, is the problem of substituting electric welding for riveting. This substitution should take place in the construction of ships, boilers, bridges, and in the industrial arts. It is a large problem but a solution of the electric welding problem will eventually very largely replace the art of riveting for patching together of steel members. In the construction of a ship's hull, smooth steel plates from $\frac{3}{8}$ in. to $\frac{3}{4}$ in. thick, 6 ft. wide, and from 15 ft. to 30 ft. long must be joined to one another and to the framework. The plates must be welded to one another in such a way that the builder can be certain that the joint is equal to or similar in strength and ductility to the present riveted construction. Experience has shown that the ordinary electric weld, using the metallic electrode process, for samples 2 in. or 3 in. wide and from $\frac{3}{8}$ in. to $\frac{3}{4}$ in. thick will produce a weld that, when made with sufficient care and tested in the pulling machine is very much stronger than the best riveted joint. The writer, however, does not think that this statement can be made when large sheets are welded

* President Lincoln Electric Co. † Received Mar. 26, 1919.

together, as in the construction of a ship's hull. The heating, which is necessarily a part of the electric welding process, introduces contraction stresses between the plates that are welded and these internal stresses, in some cases, are practically equal to the strength of the weld. It is impossible by any means now known to determine the amount of these stresses and, consequently, the use of the riveting process for welding the plates, which is much more expensive and not so strong, is universally used because it eliminates these stresses. In some cases these internal stresses can be relieved by annealing the whole sheet, including the joints, after it is welded. This process has been used with entire success for welding some tanks for holding ammonia under a pressure of 200 lb. per sq. in. It is obvious that in the construction of a ship's hull such an annealing process is out of the question. The conclusion, therefore, is that if the construction of ship's hulls by electric or gas welding is to be successful, the weld must be of such a nature that the welded material can stretch and relieve the internal or contraction stresses and this stretching must not crack the joint at any time nor weaken the strength of the welded joint.

One or two hundred tests on metallic welded joints with the best electrode that could be secured convinced the writer that the metallic electrode welding process will not produce a joint that will fulfill the above specifications; consequently, his attention has been directed toward the use of the carbon arc. He has developed the practice to a point where it is possible to say that the metal in the weld fulfills the specifications necessary to take care of the internal stresses outlined above. In other words, the practice of carbon arc welding has been advanced to the point where it is easy to make joints that are past the elastic limit of ordinary steel plate. The metal in these welds at the same time is ductile so that a test piece of metal cut from the weld itself and pulled in a testing machine will show satisfactory elongation and reduction of area. The writer hopes within a short time to be able to present the details of this process.

HERMANN LEMP,* Erie, Pa. (written discussion†).—This paper is a very good résumé of the art of fusion and spot welding of mild steel, but it has entirely overlooked the well-known process of resistance welding, originated by Prof. Elihu Thomson. The Thomson process, the sire of spot welding, is so well established in our industries and has been so thoroughly discussed in the technical press and the records of the A. I. E. E. that it has almost passed the stage of debate, but may well be recalled to the attention of the younger engineers.

When one considers the vast quantities of mild-steel products daily welded in the United States alone by that process, and this, with unskilled

* Engineer, Erie Works, General Electric Co. † Received Mar. 27, 1919.

persons operating automatic and semi-automatic machinery, with a speed unexcelled by the fusion process, the writer may be pardoned for drawing attention to one characteristic feature of the process; namely, that parts to be welded are substantially protected against contact with the oxygen of the air, and what few oxides may be formed near the edges are effectively squeezed out of the joint by the pressure exerted on completing the weld. In addition, all kinds of metals and alloys, some of which are termed unweldable, are successfully welded thereby. Crane chains, harness rings, automobile rims, bicycle tubing, wire fences, carriage tires, metal wheels, etc., are welded daily in such quantities as would astonish any one when confronted with actual figures, and in the welding of valve flanges to cylinders of Liberty aviation motors at Detroit, the Thomson process easily surpassed the gas and electric-arc processes, as regards both speed and excellence of finished product.

Spot welding was the first process considered by the Welding Committee created by the A. I. E. E. in the summer of 1917, and its adaptability to ship-plate welding was clearly recognized at that time in respect to fabricated parts that could be brought to a stationary spot welder in place of having the spot welder brought to the work. The Thomson Electric Welding Co., of Lynn, designed a portable spot welder for ½ in. ship plates. The long secondary leads required for spanning the gap, and consequently reduced output by reactance, failed to give satisfactory performance with the transformers originally designed for ½-in. plates. The increased capacity of the transformer required to overcome the effects of reactance has always been a bugaboo in connection with spot welders or electric welders of large dimensions, and the method of overcoming this reactance by making two welds in series by two transformers placed on opposite sides of the plates suggested itself to the writer in August of 1917, and was communicated to Mr. Hobart at that time. This statement in no way is meant to reflect upon the able work done by Mr. Weed, who independently developed and is responsible for the practical and successful design of the Duplex spot welder referred to in Mr. Hobart's paper. As the current flowing through both welds is substantially the same, the heating effect will depend entirely on the local resistance, which, in turn, depends on contact pressure. It is this difficulty that has been overcome in so satisfactory a manner by Mr. Weed by utilizing independent contact pressure for the two welds made simultaneously.

Spot welding, as the name implies, may be carried out in two ways. The electrodes through which the current is passed to the plates may be of so small a section that only the metal directly under the contacts or in its immediate vicinity becomes heated; or the plates to be welded together may be provided with raised surfaces, or teats, which form a restricted contact with each other and are heated to welding tempera-

ture. In the latter case, the electrodes supplying the current need not be so small in area as in the former modification of the process. As the thickness of the plates to be welded is increased, both the current density and the pressures are materially increased, and the problem of producing a satisfactory electrode with good wearing qualities becomes more difficult.

From his experience with an electric armor-plate annealer, which he designed in 1894, the writer has learned that tremendous quantities of current can be passed through relatively small contact surfaces of copper, provided they are backed by an increasing mass of copper properly water-cooled; 50,000 to 60,000 amp. were taken through contacts 1 in.

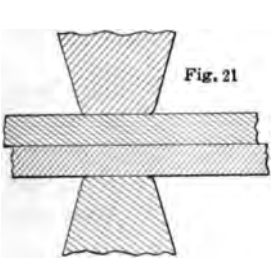

Fig. 21

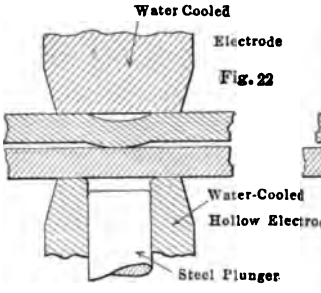

Fig. 22

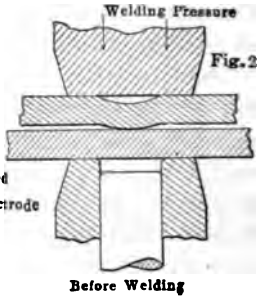

Fig. 23

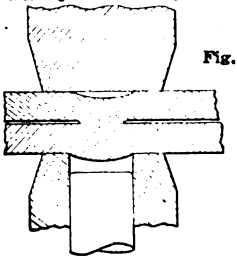

Fig. 24

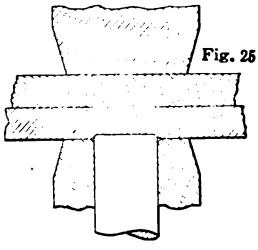

Fig. 25

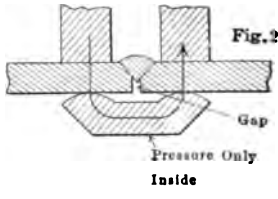

Fig. 26

square, rounded off so as to make almost a point contact at times, and these contacts would last almost indefinitely, but in that particular application the pressure on the contacts was relatively light. He has been assured by Mr. Merrill, who was connected with the heavy spot-welding work referred to in Mr. Hobart's paper, that the problem of contacts is not as serious as one may think. It may be well, however, to consider the modification shown in the accompanying sketches, when large energies for spot welding are not available.

Fig. 21 shows the present method, in which two plates are placed in contact with each other and current passed through them through contact surfaces of restricted area. In this case the pressures on the contact surfaces have to be considerable and the surfaces themselves are relatively small.

Fig. 22 shows one of the plates provided with a series of teats produced by a punch press, similar to the one used for punching holes for riveting but in which the punch is rounded and is allowed to enter a small distance into the metal. Such a plate, when placed in contact with a water-cooled electrode of liberal dimensions, will make contact only on a ring surface, not in the middle. The other plate, which is without teats, rests on a hollow electrode, also of liberal dimensions, as far as contact surface is concerned. The teat in the upper plate insures a contact with the other plate at the apex of the teat and will be heated at the point of contact with the other plate. Owing to the fact that the lower plate rests on a circular electrode unsupported and uncooled in the center, the center will be heated evenly with the teat in the upper plate. In the center of the lower electrode, but normally not in contact therewith, is a steel plunger, which may be separately operated by either hydraulic or pneumatic pressure.

Fig. 23 shows the plates in contact under ordinary pressure suitable for proper heating at the beginning of the weld. Fig. 24 shows the plates on the completion of welding; they are still under contact pressure but the hydraulic pressure has not been applied. It will be noted that the plates have been pushed together and a cup-shaped depression formed at the bottom of the lower plate that fits into the space in the hollow electrode. It will also be noted that the hollow depression in the top of the upper plate, formed at the time when the teat was made, still exists. Fig. 25 shows the last part of the process. After cessation of current, pressure on the central plunger is applied, leaving the upper plate restored to its original form with a spot weld in the center and a cup-shaped depression in the bottom.

By this method we have, first, the certainty of concentrating the heat always at the teat, even if the plates should be slightly bent; second, lasting copper contacts; and, third, the heated metal is worked, reducing its loose structure to a denser fiber. The drawbacks of the method are that one of the plates, at least, will require an additional mechanical operation to provide the teats, and once these teats are formed, welds can be made at no other place unless loose metal buttons are inserted.

Welding Ship Plates with One Transformer.—A method of welding ship plates by spot welding, which requires only one transformer in close contact with the plates, and hence a relatively low resistance, is shown in Fig. 26. The two plates to be welded are butted together without actually touching, leaving a narrow gap. A butt strap, or individual strap with teats, is then pressed against the inner side of the plates. The transformer is on the outside of the hull of the ship, so the current passes from plate to plate through the strap, and pressure is applied from the inside. After the welds are completed, the gap is closed by arc welding for calking purposes only. The welder and pressure devices, both inside

and outside of the plates, are to be pulled together by direct-current electromagnets, which on the basis of 100 lb. per sq. in. can be made of sufficient size to give the necessary pressure.

Polarity.—For carbon-arc welding, the carbon electrode should be connected to the negative terminal; if connected to the positive, there is danger of making a hard weld, which cannot be properly machined. The positive pole is always hotter than the negative, hence, when welding with metal electrodes, it is best to connect the positive to the plates as they are usually of greater volume than the electrode and should be heated sufficiently to produce fusion. Under certain conditions only, as when very thin plates are to be welded with the metallic arc, it might be well to connect the positive to the electrode. When welding with alternating currents, graphite electrodes may be used when the plates to be welded have turned up edges to be united together and where no machining is required after welding. On general principles, metallic electrodes are more suitable for alternating currents.

HARRY D. MORTON,* Detroit, Mich. (written discussion†).—In section 22, Mr. Hobart says that machinery is being developed for automatically feeding the electrode. Having been, for a long period, engaged in this work, the writer gladly takes part in this discussion with the thought that he may be able to contribute something of interest concerning the rapidly developing art of welding.

Theory of Electric Arc.—The generally accepted theory of the electric arc is that part of the electrode material is vaporized, and that this vaporous tube or column forms a path for the electric current. As a result of the vaporous character of the current path, all arcs are inherently unstable; and the maximum of instability is no doubt found in that form of arc employed for metallic-electrode welding purposes. We here have, in conjunction with the natural instability characteristic of all arcs, rapidly fusing electrode materials and the disturbing effect of the constant passage through the arc of a large quantity of molten metal to form the weld. This molten metal must pass through the arc so rapidly that it will not be injured or materially contaminated, otherwise the weld will be useless. Prima facie, the combination of these unfavorable conditions would seem to justify fully the skepticism of most electrical engineers as to the possibility of effecting such control of the metallic arc as to permit of uniformity and continuity in welding results. In addition, there is another and more important factor, and one that seriously mitigates against this desired uniformity and continuity, namely, the personal equation of the operator. The consensus of opinion, so far as is known to the writer, seems to be that about 95 per cent. of the welding result is dependent on the skill of the operator and that at least six months' practice is necessary to acquire reasonably satisfactory proficiency.

* Secretary-Treasurer, Automatic Arc Welding Co. † Received Mar. 27, 1919.

As the result of thousands of observations of welds produced automatically (wherein the personal equation is entirely eliminated), the writer inclines toward the theory that the molten electrode material passes through the arc in the form of globules; and that where ⅛-in. electrode material is employed with a current of about 150 amp. these globules are deposited at the rate of approximately two per second

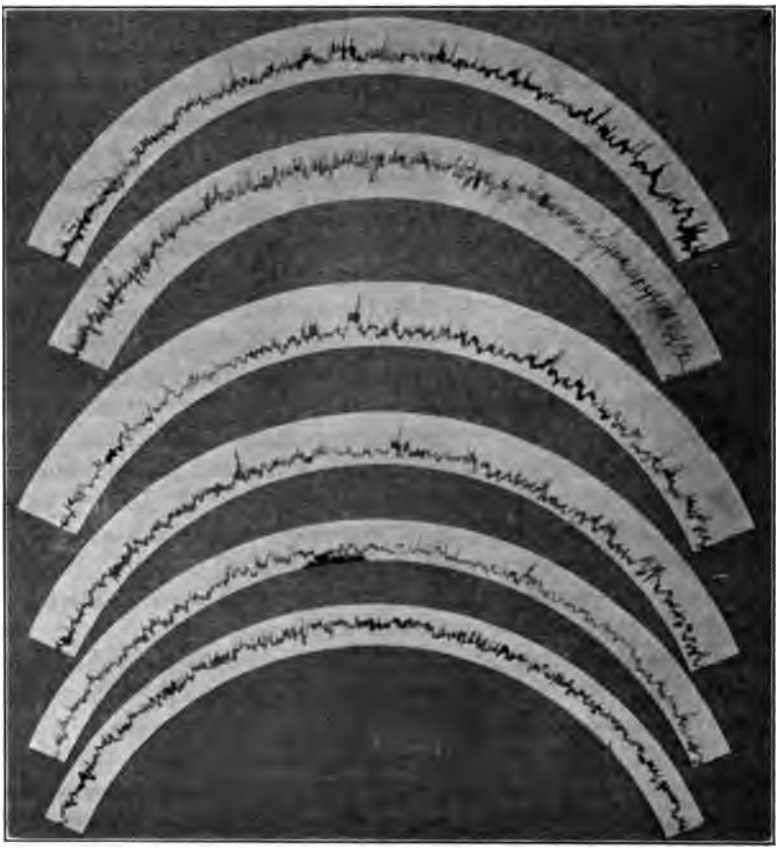

FIG. 27.—TYPICAL AMMETER CHARTS OF OPERATION OF MORTON AUTOMATIC, METALLIC-ELECTRODE, ARC-WELDING MACHINE. AVERAGE TIME ABOUT 1 MIN. 45 SEC.

The passage through the arc of each globule apparently constitutes a specific cause of instability in addition to those existent with slowly consumed electrodes. This hypothesis seems to be borne out by ammeter records (typical specimens of which appear in Fig. 27), together with the fact that the electrode fuses at the rate of about 0.2 in. per sec. Moreover, the globules appear to be approximately equal in volume to a piece of wire 0.125 in. in diameter and 0.1 in. long.

Essentials for Maintaining Uniform Length of Arc.—Assuming this theory to be correct, to maintain a uniform arc length in manual weld-

ing, the operator must feed the electrode toward the work at the rate of 0.1 in. upon the deposition of each globule, in other words, 0.1 in. twice per second, a synchronism beyond human attainment. Simultaneously with such feeding, the arc must be moved over the work to melt the work material, distribute the molten electrode material, and form the weld. Inasmuch as the effect of the arc is highly localized, it is reasonable to suppose that different parts of the welding area present relatively wide variations in respect to temperature, fluidity, and conductivity of the molten mass—controlling factors not within the ken of the human mind. The situation is further complicated by the facts that neither the welding wire nor the work material is uniform in fusibility or in conductivity, and that the contour of the work varies continually as its surface is fused and the molten metal is caused to flow. The belief is general that a very short arc is productive of the best welding results; but it is an arc of this character that makes the greatest demands on the skill of the operator, for there is always the danger that the electrode will actually contact with the work and destroy the arc.

Effect of Fluctuations of Arc Length on Fusing Energy.—As the fusing energy of the arc varies widely with fluctuations in the arc length and as the uniformity of the weld depends on the constancy and correctness of this fusing energy, it seems remarkable that operators are able ever to acquire such a degree of skill as to enable them to produce welds that are even commercially satisfactory. Further, so far as the writer is informed, there is no means, other than such as would be destructive, for determining whether a completed weld is good or bad. The logical solution appeared, to the writer, to be the elimination of the personal equation and the substitution therefor of means whereby tendencies toward variations in the arc would be caused automatically to correct themselves, just as a steam engine, through the action of its governor, is caused to control its own speed.

Methods of Mechanically Stabilizing and Controlling the Arc.—The efforts of the writer, for a number of years, have been directed toward stabilizing and controlling the metallic arc, and applying such stabilizing and controlling means to two general lines of welding machinery: (1) machines wherein the feeding of the electrode wire, with reference to the work, and the producing simultaneously therewith of relative movement between the wire and the work are both automatically effected; and (2) what, for lack of a better term, might be called a semi-automatic machine, in which the feeding of the electrode and the control of the arc are accomplished automatically but the traversing of the electrode with reference to the work is manually effected by the operator, permitting him the exercise of judgment with reference to the quantity of metal to be deposited in various parts of the groove. The automatic machine has been in successful operation for a long period and the semi-

automatic machine for about five months. While the goal was not attained without many difficulties and a great expenditure of time and money, the results have been surprisingly successful.

Because of the lack of any definite data as to what actually occurs in this form of arc, or why it occurs, due, no doubt, to the impossibility of differentiating between phenomena that are characteristic of the arc and phenomena due to the personal equation of the welder, it seemed logical that the initial step should be to so environ the arc that it would not be subject to erratic extraneous influences, to the end that reasonably definite determinations might be substituted for scientific speculation. In the design and construction of the machines, great care was exercised to minimize the possibility of mechanical defects that might lead to erroneous conclusions. Starting with the assumption that the work could only be based on open-minded observation of the behavior of the arc under machine control, an automatic welding machine was built in which was incorporated the greatest possible number of adjustable features, in order that, if necessary, it might be possible to wander far afield in the investigations. This adjustability has proved invaluable in that it has permitted logical, consistent, and sequential experimenting over a very wide range of conditions. Working under these favorable circumstances, there were soon segregated a few clearly demonstrable facts to serve as a foundation for the structure, which has since been added to, brick by brick, as it were.

Efforts have been directed toward the practical rather than the scientific aspect of the subject. The operation of the automatic machines has brought to light many curious and interesting phenomena, some of which appear to negative conclusions heretofore formed which have been predicated upon observations made in connection with manual welding. It is hoped that these and other phenomena, which can thus be identified as purely arc characteristics, will be the subject of profitable scientific investigation when time is available for this purpose.

Development of Automatic Machines.—In the five forms of machines made in the course of the development, the welding wire is automatically fed to the arc; and, in the first four machines, the relative movement between the work and the welding wire is automatically and simultaneously effected. Early in his investigations, the writer concluded that a substantial equilibrium must be maintained between the fusing energy of the arc and the feeding rate of the welding strip; and it soon became evident that if the welding strip is mechanically fed forward at a uniform rate equal to the average rate of consumption with the selected arc energy, this equilibrium is actually maintained by the arc itself, which seems to have, within certain circumscribed limits, a compensatory action as follows: When the arc shortens, the resistance decreases and the current rises. This rise in current causes the welding strip to fuse more rapidly

than it is fed, thereby causing the arc to lengthen. Conversely, when the arc lengthens, the resistance increases, the current falls, the welding strip is fused more slowly than it is fed, and the moving strip restores the arc to its normal length.

FIG. 28.—PILOTED CUP AUTOMATICALLY WELDED BY METALLIC-ELECTRODE ARC PROCESS TO TUBE TO FORM 75-MM. SHRAPNEL SHELL. ANALYSIS OF ELECTRODE MATERIAL: SILICON, 0.02 PER CENT.; SULFUR, 0.013 PER CENT.; PHOSPHORUS, 0.07 PER CENT.; MANGANESE, TRACE; CARBON, 0.07 PER CENT.; ALUMINUM, 0.038 PER CENT.

While this compensatory action of the arc will maintain the necessary equilibrium between the fusing energy and the feeding rate under very carefully adjusted conditions, this takes place only within relatively

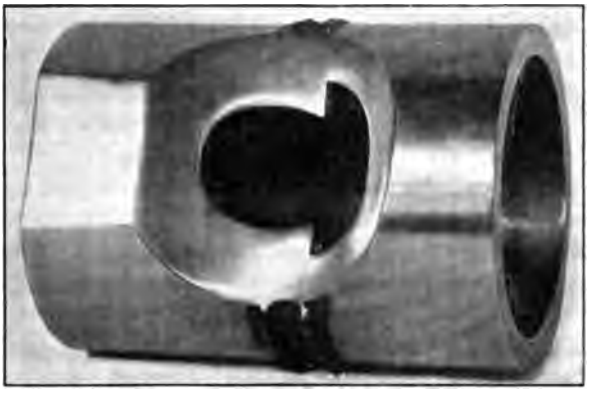

FIG. 29.—PILOTED CUP AUTOMATICALLY WELDED BY METALLIC-ELECTRODE ARC PROCESS TO TUBE TO FORM 75-MM. SHRAPNEL SHELL. ANALYSIS OF ELECTRODE MATERIAL: SILICON, 0.03 PER CENT.; SULFUR, 0.049 PER CENT.; PHOSPHORUS, 0.008 PER CENT.; MANGANESE, 0.31 PER CENT.; CARBON, 0.28 PER CENT.

narrow limits. It was very apparent that, due to variations in the contour of the work, and, perhaps, to differences in the fusibility or conductivity of the welding strip or of the work, the range of this self-com-

pensatory action of the arc was frequently insufficient to prevent either contacting of the welding strip with the work or a rupture of the arc due to its becoming too long. The problem that arose was to devise means whereby the natural self-compensatory action of the arc could be so greatly accentuated as to preclude, within wide limits, the occurrence of marked arc abnormalities. There was ultimately evolved, by experiment, such a relation between the fusing energy of the arc and the feeding rate of the welding strip as to give the desired arc length under normal conditions; and tendencies toward abnormalities in arc conditions, no matter how produced, were caused to bring into operation compensatory means for automatically, progressively, and correctively varying this relation between fusing energy and feeding rate, such compensatory means being under the control of a dominant characteristic of the arc. In

FIG. 30.—PILOTED CUP AUTOMATICALLY WELDED BY METALLIC-ELECTRODE ARC PROCESS TO TUBE TO FORM 75-MM. SHRAPNEL SHELL. ANALYSIS OF ELECTRODE MATERIAL: SILICON, 0.02 PER CENT.; SULFUR, 0.032 PER CENT.; PHOSPHORUS, 0.008 PER CENT.; MANGANESE, 0.20 PER CENT.; CARBON, 0.18 PER CENT.

their ultimate forms, the devices for effecting the control of the arc are simple and entirely positive in action, making discrepancies between fusing energy and feeding rate self-compensatory throughout widely varying welding conditions. For instance, the shrapnel shell shown in Fig. 28 was automatically welded with wire differing greatly in chemical constitution from that used on the shell shown in Fig. 29 (see analyses), yet no change was made in either the mechanical or the electrical adjustments. The radically different welding conditions were compensated for solely by the operation of the automatic control. The electrode materials used for the shells shown in Figs. 30 and 31 differed so greatly from those employed respectively in welding the shells shown in Figs. 28 and 29 that a change in the relation between fusing energy and feeding rate had to be made manually. After this adjustment was made, the shells were welded with their respective electrodes, which varied widely in their chemical

constitution, without further manually changing either the mechanical or the electrical conditions.

FIG. 31.—PILOTED CUP AUTOMATICALLY WELDED BY METALLIC-ELECTRODE ARC PROCESS TO TUBE TO FORM 75-MM. SHRAPNEL SHELL. ANALYSIS OF ELECTRODE MATERIAL: SILICON, 0 04 PER CENT.; SULFUR, 0 016 PER CENT.; PHOSPHORUS, 0.058 PER CENT.; MANGANESE, NONE; CARBON, 0.24 PER CENT.

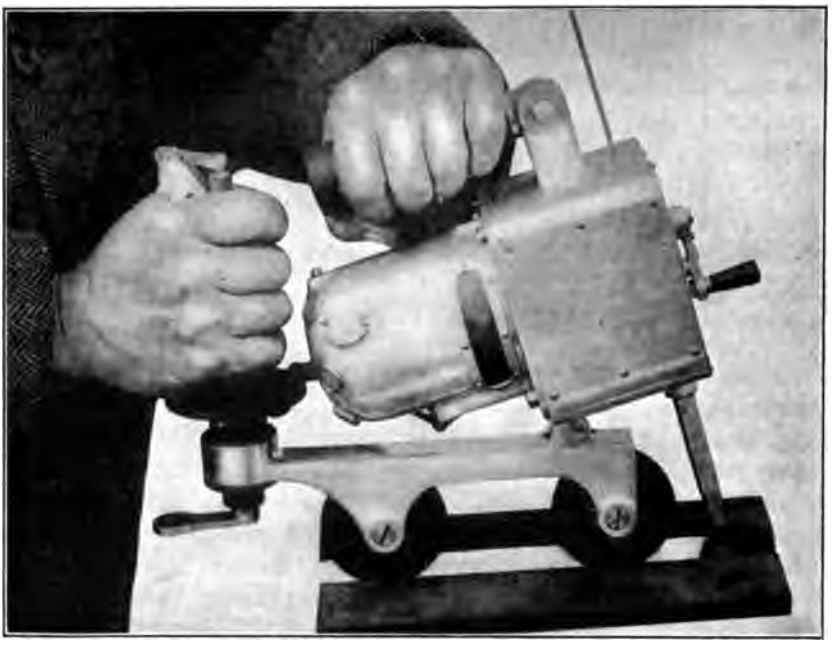

FIG. 32.—MORTON SEMI-AUTOMATIC METALLIC-ELECTRODE ARC-WELDING MACHINE. THE ELECTRODE IS AUTOMATICALLY FED TO THE ARC, WHICH IS AUTOMATICALLY MAINTAINED, WHILE THE MACHINE IS MANUALLY MOVED ALONG THE GROOVE TO BE WELDED.

In a recent test of the semi-automatic machine, successful welds were made under the condition that the impressed voltage of the weld-

ing generator was changed throughout a range of from 50 to 65 volts, without necessitating any manual adjustment. The only observable effects of the wide variations in the supply voltage were slight differences in the arc length.

In short, the compensatory action of the control has proved effective over so wide a range of welding conditions, not only as to the electrical supply and chemical constitution of both electrode and work materials, but as to extensive variations in the contour of the work and in many other particulars, as to make it seem apparent that the machines do not represent merely successful laboratory experiments but are suited to the requirements of actual commercial welding.

Angle of Electrode to Work and Insertion of Extra Resistance.—One particularly interesting observation resulting from the experiments is that the angle of inclination of the electrode with reference to the work is very important. An angular variation of 5° will sometimes determine the difference between success and failure in a weld. About 15° from the perpendicular works well in many cases. In welding some materials, the electrode should drag, that is, point toward the part already welded rather than toward the unwelded part of the seam.

While it has been customary in some welding systems to provide means whereby extra resistance is inserted in series with the arc at the instant of the initial contact which starts the flow of current, which resistance is automatically cut out upon the striking of the arc, experience with the automatic machines indicates that this is quite unnecessary.

Relation of Electrode to Work Materials.—Early in the experiments, it was noted that in many cases there was a decidedly marked affinity between particular electrode materials and particular work materials. A slight change in either element affects the degree of this affinity. While it has invariably been possible to control and maintain the arc and weld continuously, in some instances incompatibility between electrode material and work material has been productive of interesting phenomena. For instance, the combination of work material (steel of about 0.45 per cent. carbon content) and the particular electrode material shown in Fig. 28 produced an arc that was remarkably quiet and free from sputtering. Throughout the weld, this arc was suggestive of the quiet flame of a candle or lamp, the erratic behavior that we are accustomed to associate with the ordinary metallic arc being absent. The effect is reflected in the uniform deposition of the welding material.

On some classes of work material Bessemer wire, which some authorities claim cannot be used in metallic-electrode arc welding, produces an arc and a weld very satisfactory in appearance. On other work material, the Bessemer wire arc is violently explosive. These explosions are accompanied by quite sharp reports and the scattering over some considerable distance of globules of molten metal frequently $\frac{3}{32}$ in. or more

in diameter. Under certain other conditions, apparently growing out of incompatibility between the work material and the electrode material, the oxygen flame accompanying the arc gyrates very rapidly about the arc, producing an effect suggestive of the "whirling dervish."

From both the practical and the scientific points of view, the writer has experimented quite extensively with varying combinations of work material and electrode material. Throughout all the differences in arc conditions, many of which palpably accentuate the natural inclination toward instability, the control has so operated as to justify the expression "the arc persists."

Generally speaking, the Swedish and Norway iron wires seem to produce more quiet arcs and, possibly, a more uniform deposition of electrode material, than do wires of other classes. These welds may perhaps be found to be slightly more ductile than those made with wires

FIG. 33.—TUNGSTEN HIGH-SPEED RING AUTOMATICALLY WELDED BY METALLIC-ELECTRODE ARC PROCESS TO COLD-ROLLED CORE TO FORM MILLING-CUTTER BLANK.

of other chemical composition. On the other hand, these soft wires, although undoubtedly of relatively high fusibility, do not, for some reason, seem to produce an arc that cuts into some work material as deeply as might be desired, nor as deeply as do the arcs formed with certain other kinds of wire. Considered from every angle, the writer is disposed to regard the Roebling welding wire as the best he has thus far tested for use on mild steel. This wire produces a reasonably quiet arc which seems to cut into the work to more than the ordinary depth, while, at the same time, the electrode material is fused with more than average rapidity—thus increasing the welding rate.

While scientists will no doubt ultimately arrive at the correct hypothesis for solving the problem of why one combination of electrode material and work material is productive of better results than can be obtained with another combination, the writer's conclusion is that, with

the data at present available, the determinations must be made by actual experimenting—having in mind the qualities desired in the particular weld, such as ductility, tensile strength, elongation, and elastic limit. Inasmuch as it is possible, with the automatic machine, to maintain arc uniformity with practically any kind of electrode material and to produce welds which, under low magnification, at least, appear to be perfect, and which respond favorably to ordinary tests such as bending, cutting and filing, it is reasonable to conclude that proper selection of electrode material will be productive of perfect welds on any kind of work material. To date, no steel has been tested on which apparently satisfactory welds could not be made. High-speed tungsten steel has been successfully welded to cold-rolled shafting, using Bessemer wire as electrode material, as is shown in Fig. 33. Ordinary steels varying in carbon content from perhaps 0.10 to 0.55 per cent. have been welded with entire success.

Condition of Metal in Crater.—Because of the fact that the complete welding operation has been automatic and may be continued for a considerable length of time, say 5 min., an exceptional opportunity has been afforded for close concentration upon and study of the appearance of the arc. What seems to occur is that the molten metal in the crater is in a state of violent surging, suggestive of a small lake lashed by a terrific storm. The waves are dashed against the sides of the crater, where the molten metal of which they are composed quickly solidifies. The surgings do not seem to synchronize with nor to be caused by the falling of the globules of molten metal into the crater, but seem rather to be continuous. They give the impression that the molten metal is subjected to an action arising from the disturbance of some powerful force associated with the arc—such, for instance, as might result from the violent distortion of a strong magnetic field. Altogether, the crater phenomena are very impressive; and the writer hopes ere long to be able to have motion pictures made which, when enlarged, should not only afford material for most fascinating study, but also throw light upon some of the mysterious happenings in the arc.

Electrodes and Currents Used.—So far, electrode wires $\frac{1}{8}$ in. in diameter have been chiefly used in the machines. Successful welds have been made with current values ranging from below 90 to above 200 amp. at impressed voltages of 40, 45, 50, 55, 60, 65, and 80. Under these varying conditions, the voltage across the arc has been roughly from 16 to 22. The machines have thus far been run only on direct current. Inasmuch as it is possible, by electrical and mechanical adjustments, to establish nearly any arc length that may be found to be most desirable for a particular class of work, and as the control system will maintain substantially that arc length indefinitely, the fully automatic type of machine is nearly as certain in operation as a lathe, drill press, or other machine tool.

Automatic and Semi-automatic Welding Machine.—Upon the writer's appointment to membership on the Welding Committee of the Emergency Fleet Corpn., Professor Adams and other members of the Committee urged upon him the advisability, for purposes of ship welding, of incorporating in a small and readily portable welding tool the features of automatic wire feed and arc control. The tool designed is shown in Fig. 32. In welding with this so-called semi-automatic machine, which weighs about 10¼ lb., the operator draws the tool along the groove to be welded at such a rate as will result in the deposition of the quantity of metal required to satisfactorily effect the weld. This tool is intended for use in the many restricted spaces encountered in ship welding, which would be relatively inaccessible to a fully automatic machine. In its use, the skill required by the operator is reduced to a minimum. After having practised with the welding tool for not more than two hours, the writer demonstrated it to Mr. E. L. Hirt, of Bethlehem Shipbuilding Corpn. Mr. Hirt expressed it as his opinion that it would require six months to train a welder to such a degree of proficiency as to enable him to make a weld equally good in appearance.

Mr. Hobart, on page 572, says "There is always 0.1 in. (2.5 mm.) or more between the end of the welding rod and the work." While undoubtedly it is difficult, if not impossible, to maintain in manual welding an arc shorter than this, the writer has frequently, with the automatic machines, made continuous and strikingly good welds with much smaller arcs. In fact, in some cases there has been continuously maintained an arc so short that there hardly seemed to be any actual separation. The writer has even wondered whether, under these conditions, there was not a close approach to casting with a continuous stream of fluid metal acting as the current conveyor in lieu of or in parallel with the usually assumed vapor path. The work that has been done indicates that under automatic control much shorter arcs can be utilized than have hitherto been deemed possible, and with probable marked gain in quality of work in some instances; also, that there is much to be learned as to the mode of current action and current conduction in such an arc.

Results Obtained with Automatic Welding Machine.—With the automatic machine, black drawing steel 0.109 in. thick has been welded at the rate of 22 in. per min. A Detroit manufacturer welded manually with oxyacetylene at the rate of four per hour a large number of mine floats 10 in. in diameter, made of this material. The automatic machine made the welds at the rate of forty per hour. Liberty motor valve cages 2¾ in. in diameter have been welded to cylinders in 36 sec., as against about 5 min. required for manual welding. No. 11 gage steel tubing, shown in Fig. 34, has been welded, with an unnecessarily heavy deposit of metal, at the rate of 1 ft. per min. The productive capacity of the machines so far made has been from three to ten times that of manual welding methods,

depending on the thickness of the work material; the difference in favor of automatic welding varies inversely as such thickness. The writer is now designing an improved type of machine for use especially on heavy work, with which machine it is expected to be able automatically to lap-weld ½-in. ship plates, in the manner shown in Fig. 35, at the rate of 15 ft. per hr. One of the largest shipbuilding concerns in the United States

FIG. 34.—No. 11 GAGE STEEL TUBING AUTOMATICALLY WELDED BY METALLIC-ELECTRODE ARC PROCESS AT THE RATE OF 1 FT. PER MIN.

reports that the general average of all its manual welders on this class of work is from 1 ft. to 18 in. per hr. Other specimens of automatic welding on ship plates are shown in Figs. 36 and 37.

Bare wire only has been used in the automatic machines; and the results obtained seem to indicate that the covering of the electrodes is an

FIG. 35.—TWO ½-IN. SHIP PLATES AUTOMATICALLY WELDED BY METALLIC-ELECTRODE ARC PROCESS TO FORM LAP JOINT.

expensive superfluity. If the chief advantage of the covered electrode lies in the ability of the operator to maintain a very short arc, an arc equally short and possibly shorter can be continuously maintained by the automatic machine using bare electrodes.

No attempt has thus far been made to use the automatic machines on overhead work. The welds made with the fully automatic machine

have been of three kinds: the usual longitudinal form, annular about a horizontal axis, and annular about a vertical axis. The experience gained indicates that there is no difficulty in dealing with any ordinary contour.

As far as the maintenance of arc uniformity and e apparent character of the welds are concerned, they do not seem to be greatly influenced by

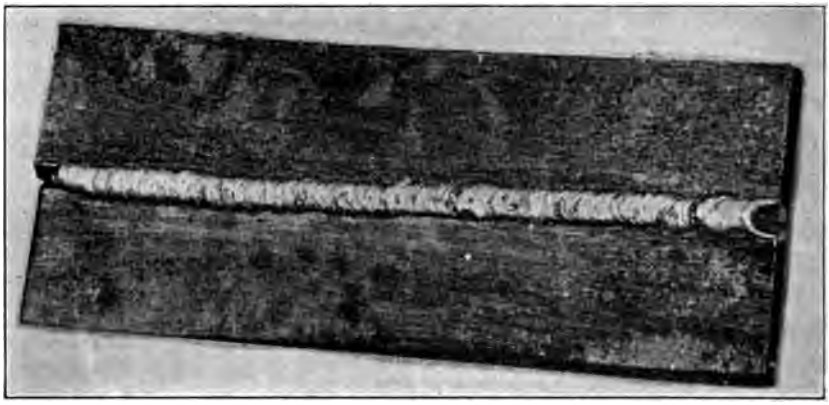

FIG. 36.—TWO ½-IN. SHIP PLATES AUTOMATICALLY WELDED BY METALLIC-ELECTRODE PROCESS TO FORM BUTT JOINT

any of the considerations referred to in the specifications appearing on page 580. The writer has repeatedly welded with wire showing evidence of pipes and seams, as well as with rusty wire and with wire covered with dirt and grease. In this connection it may be said that no pains is ever taken to remove rust, scale, or slag from the work material—

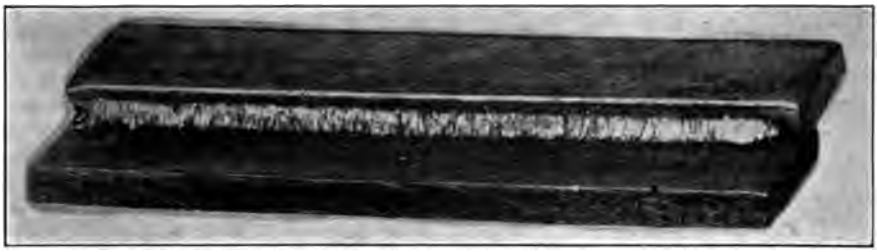

FIG. 37.—TWO ½-IN. SHIP PLATES AUTOMATICALLY WELDED BY METALLIC-ELECTRODE ARC PROCESS, SHOWING FIRST OF THREE LAYERS TO FORM LAP JOINT.

even where welds are superimposed. Apparently under uniform conditions of work traverse, arc length, and electrode angle of inclination, such as are possible in the automatic machine, impurities vanish before the portion of the work on which they occur reaches the welding area of the arc.

Referring to section 15, on the ductility of arc welds, the writer

is fully convinced that with the use of the automatic machine, ductility, like other physical properties in the weld, can be controlled by proper selection of electrode wire, in conjunction with electrical and mechanical adjustments best suited to the particular purpose in view. Automatic welds have repeatedly been made on $\frac{5}{16}$-in. mild steel which, when subjected to a 90° bend, showed a marked extrusion of the welded material but no sign of fracture. When the welded pieces are cut with a hack saw, it is very unusual to be able to note any difference in cutting qualities between the unwelded and the welded parts.

Welding Thin Material.—While the automatic machine has not been used on metal less than 0.109 in. thick, it is fair to presume that, with proper adjustments, entirely satisfactory results can be obtained on much thinner work—particularly if the nature of the work is such as to permit the use of a chill. The best method in welding very light metal seems to be to use a small electrode, a relatively low current, and a high rate of work traverse. In this way welding conditions may be controlled to almost any desired extent, because the heating action of the arc can be modified, its effect intensely localized, and the edges to be welded subjected to the fusing action for as brief a time as might be found necessary to prevent burning of the metal. These conditions, which seem to the writer to be requisite in order to weld very thin material successfully, cannot be met by the manual welder. It is here that the deficiencies incident to the personal equation become most apparent. A very slight variation in arc length or the least hesitancy in moving the arc over the work will almost certainly result in its being burned through. In short, this class of welding calls for a coördination of faculties and a delicacy of manipulation beyond the capabilities of the most skillful manual electric welder. Therefore this work is usually done with the oxyacetylene flame, wherein fusing conditions are far more easily controlled than is possible in manual metallic electrode arc welding.

R. E. WAGNER, Pittsfield, Mass. (written discussion*).—The practice of arc welding, as it exists today, consists largely of the results of experiments and practical experience; the study of the underlying scientific principles has not kept pace with its practical application. This situation, however, does not differ materially from that of many other arts where practice leads the way to the laws and principles that govern their operations.

Arc welding has been widely used for a number of years and, where properly employed, is a reliable and economical process by which steel parts may be fused together. Excellent welding is being done due largely to the skill of the operator.

Many systems have been developed, all based on some particular

* Received Apr. 1, 1919.

feature that the inventor has considered as being important. But very little was done to bring out the advantages of these systems until the Welding Committee of the Emergency Fleet Corpn. began its work of research; then it was found that few of the fundamental principles of the art were known. It was generally admitted that a short arc should be used, that the size of electrodes and amount of current required would change for different thicknesses of pieces to be welded, that pieces would have different preparations for welding depending on their thickness and strength of weld desired, etc. But the best values

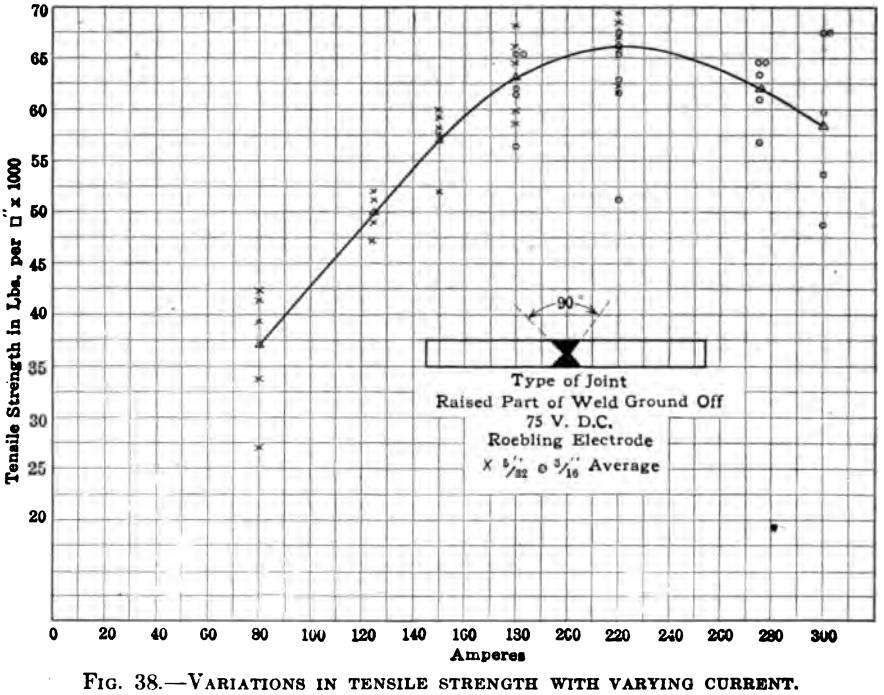

FIG. 38.—VARIATIONS IN TENSILE STRENGTH WITH VARYING CURRENT.

in any of these cases could not be definitely specified. It is, therefore, necessary that a great deal of research work be done in order to establish the fundamentals of the art.

Among the fundamentals to be investigated, for example, are: (1) How is the character of the weld affected by varying values of current with a constant voltage supply? (2) How is the character of the weld affected by varying values of supply voltage at constant current? In determining these conditions we must assume a thickness and kind of steel plate, size and kind of electrode, a preparation of plate edges for welding, a method of welding, a good operator, and a source of current.

The following data are given showing results of tests made to investigate these two questions:

In an effort to answer these questions, tests were made with ½-in. tank-steel plates that contained 0.25 per cent. carbon, 0.40 per cent. manganese, 0.025 per cent. phosphorus, 0.000 per cent. silicon, and 0.028 per cent. sulfur. The electrodes used contained 0.185 per cent. carbon, 0.561 per cent. manganese, 0.037 per cent. phosphorus, a trace of silicon, and 0.038 per cent. sulfur. The plate edges were prepared with a double *V* and had 45° and 30° bevels; they were 10 in. by 4 in. in size and the length of the weld was 10 in. Two sizes of Roebling electrodes were used, the 5/32-in. and the 3/16-in. All the welds were made by the same operator, who was a good welder; direct current was used. Later, the raised part of the welds was ground off even with the plate surface and the welded plates were cut into strips 1 in. wide for the tensile and bending tests.

With varying current and a constant 75-volt direct current, the following average values were obtained:

Current Amperes	Tensile Strength, Lb. per Sq. In.	Brinell Hardness Value	
		Plate	Weld
80	37,000	103	114
125	50,000	114	114
150	57,000	122	122
180	63,000	122	114
220	66,000	114	114
275	62,000	114	114
300	58,000	122	114

The curve given in Fig. 38 shows the plotting of values obtained in the test. The 5/32-in. electrode was used up to 220 amp. where it became so difficult to manipulate that a change was made to 3/16-in. The tests at 180 and 220 amp. were repeated with the 3/16-in. electrode in place of the 5/32-in. so that the observations would overlap and then the test was carried on to 300 amp. This curve indicates that the highest average tensile strength, in a weld of this particular kind and under the conditions specified, will be obtained with a 5/32- or 3/16-in. electrode at 220 amp. Fig. 39 shows the broken ends of the same welds covered by the curve. Better fusion conditions may be readily observed by the gradual disappearance of the vertex of the *V*. At 220 amp., the line of the *V* has disappeared. Reduction in tensile strength beyond 220 amp. is due apparently to the increased porosity of the deposited metal and difficulty in manipulation.

The curve given in Fig. 40 shows that the tensile strength is little affected by the change in voltage. The only advantage of a higher voltage is to produce a more steady current, as slight variations in the length of the arc produced by the operator will affect the total resistance

of the circuit less as the resistance of the circuit increases with the higher voltages.

The results of tests of welds made at a constant direct current of 150 amp. and voltage varying from 40 to 225 are as follows:

Current, Volts	Tensile Strength, Lb. per Sq. In.	Current, Volts	Tensile Strength, Lb. per Sq. In.
40	50,000	150	57,000
75	57,000	200	60,000
100	54,000	225	58,000
125	57,000		

Fig. 41 shows the variations in current at low and high voltages during welding operations. The higher voltage gives the more steady current. Various systems have been invented to maintain a constant current in the arc with a low supply voltage, but a good operator will not produce sufficient variations in his current to materially affect the strength of his weld at any voltage down to 40 volts direct current.

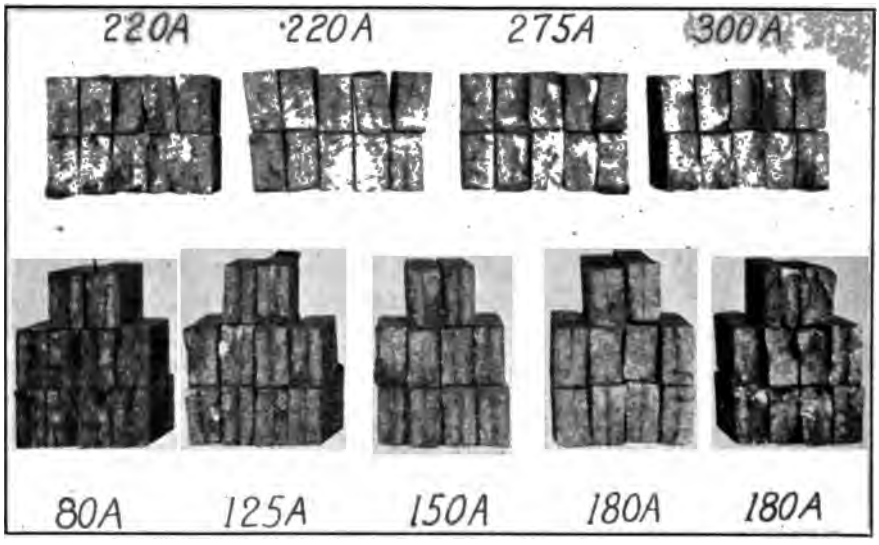

FIG. 39.—APPEARANCE OF WELDS WHEN TEST PIECES WERE BROKEN.

These observations indicate that for any particular condition, the selection of the proper current is important if the strongest weld is to be obtained; and that if the current is right, the voltage of the supply system if it is between 40 and 225 may be disregarded.

Bending tests were made for each condition of current and voltage, but it was not observed that the bending characteristics of the welds were in any way altered by the variations in current and voltage noted.

These observations are only a very small part of the broad field that is open for investigation and research. It is hoped that many

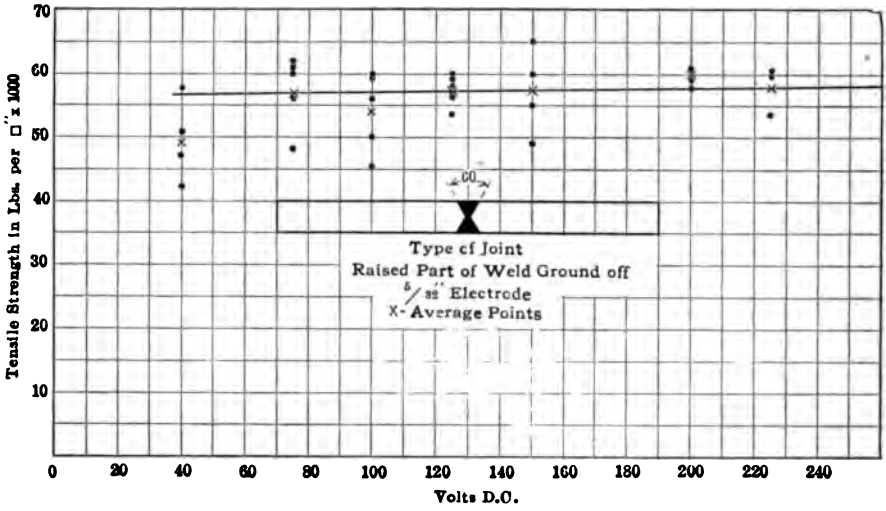

FIG. 40.—VARIATIONS IN TENSILE STRENGTH WITH VARYING VOLTAGE, AMPERES CONSTANT AT 150.

investigators will submit data bearing on these two and other essential features of welding.

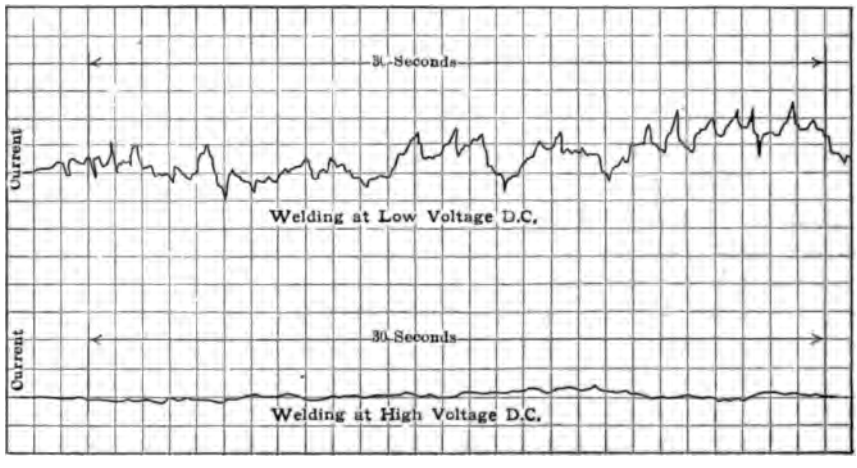

FIG. 41.—COMPARATIVE STABILITIES OF CURRENT VALUE DURING WELDING AT LOW AND HIGH VOLTAGES DIRECT CURRENT.

T. T. HEATON,* Uxbridge, Middlesex, Eng. (written discussion†).—For the past 27 years the writer has been using electric arc welding with

* The Steel Barrel Co., Ltd. † Received Mar. 31, 1919.

a carbon electrode. If metal is to be added to the weld, it is laid over the seam and fused by the direct action of the arc. But in many cases no metal is added, the parts to be united are simply joined together by fusing. A good deal of importance must be attached to the design of the parts to be welded so as to get good results in the welds and to avoid, as far as possible, any distortion due to the effect of contraction.

On page 582, Mr. Wagner is quoted as saying that it is very difficult to weld with a carbon arc when the polarity of the carbon is positive. In welding mild steel or iron, the writer has found this to be quite true. For various reasons it is best to weld with the work positive and the carbon negative. Any deposit from one pole to the other takes place from the positive to the negative; therefore it is not possible to carry any foreign matter into the weld, so no carbon from the electrode can be mixed with the weld. Besides, the positive pole is considerably hotter than the negative, which is an advantage. It was stated recently that the temperature of the electric arc is approximately 3650° or more C., and that the positive pole is about 1000° in excess of the negative.

On page 594, certain criticisms are given: "(*a*) Carbon is carried into the deposited material, thus reducing its ductility." With the arrangement of poles in the manner adopted by us this does not occur. Analyses show that the result of the welding is to remove more or less all impurities from mild steel, leaving a very considerably increased proportion of pure iron, as compared with the metal before welding. This applies to welds made in sheet steel where added metal was simply scrap of the same material.

"(*b*) It is more difficult to obtain good fusion." I do not find that this is so.

"(*c*) It is more difficult to manipulate" than the metallic electrode system. This is true to a certain extent, but so far as our experiences go, the carbon-arc welds show no disadvantage in comparison. Probably this is due to the fact that the operator with the carbon arc possessed, in the tests made, greater ability through long practice. The shorter training of many operators with the metallic arc may account for this. But I consider that there is better fusion with the carbon electrode as the operator can distribute the arc as long as he requires to produce a good weld and to direct it in any way that may be suitable, whereas with the metal electrode, he is limited to the time required to melt it.

With regard to criticism (*d*) we do not find that there is any discomfort to the operator.

"(*e*) Greater cooling stresses are developed because larger areas of adjacent metal are heated." For some classes of joints it is an advantage to spread the heat over a fairly wide area, as the metal surrounding the weld is improved thereby.

At the end of the page, certain counter-statements are made: "(*a*)

No preparation of the abutting edges is necessary." I scarcely agree with this unless the metal is thin.

"(*b*) Greater rate of deposition of metal and greater speed of welding" is possible. This again depends on the thickness of metal.

"(*c*) Probable greater adaptability to automatic welding." So far as my experience goes, attempts at automatic welding with the carbon electrode have not been successful and in any case can only be limited to very special examples.

There is no doubt whatever that electric welding by this method is exceedingly successful, both technically and commercially, but our experience is that it is best to specialize the operators in certain kinds of work, as it is very difficult to find welders capable of doing a variety of work with success. To do this requires not only practice but a high order of intelligence in the workman.

Some time ago, we made some experiments in the speed of welding by this system with the following results:

A straight weld uniting two sheets 3½ mm. thick took 2¾ min. per ft. run to make a perfectly sound hammered weld. A stamped steel boss was welded into a mild steel sheet, also 3½ mm. thick. The circumference of weld, 17¼ in., was made first by welding one side and turning over and welding the opposite side. The outside weld occupied 2½ min. and the inside 2 min. These are, of course, good times and the welds were made by a practised workman.

The writer has followed the work done by the Welding Research Sub-committee in America with very great interest and is of the opinion that its labor will confer a benefit to engineering science.

CHAS. R. DARLING,* London, Eng. (written discussion†).—It may be of interest, in connection with Mr. Hobart's paper, to indicate the present position of electric welding in Great Britain. In general, the progress of spot welding is somewhat slow, but the development of arc welding has been rapid during the past two years. Among the larger structures to which arc welding has been applied are: a rivetless barge with a displacement of 225 tons that has given excellent service during the last nine months, pontoons, and experimental ships under construction by Messrs. Cammell, Laird & Co., at Birkenhead, and Messrs. Swan, Hunter, & Wigham Richardson, at Newcastle. Details of the last, with respect to cost of production and seaworthiness, are awaited with interest. In shipyard and general engineering repairs, arc welding is growing rapidly in popularity. Mention may be made of the successful deposition of manganese steel on mild steel on the rims of dredger buckets and in other cases where a hard wearing surface is required.

All who have experimented, in this country, are agreed that a thorough

* Chairman, British Electric Welding Research Committee.

† Received Apr. 2, 1919

investigation of the destructive iron arc is necessary in order to arrive at the best method of working. At present, it is not possible to insure that two consecutive welds, made under apparently identical conditions, will give even approximately similar results under test. An inspection of the Wirt-Jones tests, on page 608, makes this clear; in my own experience it has been quite common to obtain results varying between 62,000 and 52,000 lb. in ultimate tensile strength, and 8.0 to 4.5 per cent. of elongation, without any obvious reason for the difference. Until we know what is going on in the arc during the process of welding, it is unlikely that a standard method of working can be formulated.

One notable difference in the practice in America and in this country is the general adoption of the coated, or fluxed, electrode by British welders, in contrast with the wide use of bare electrodes in America. Workmen claim that it is much easier to work with a fluxed electrode, as it is less difficult to maintain the arc; also that a better lateral spreading of the deposit is secured by the use of flux. It is difficult to overcome the prejudice of the welders, so most employers use fluxed electrodes in spite of the greatly increased cost. My impression is that much of the trouble experienced with bare electrodes has been due to improper chemical composition, and might be overcome by the use of rods of uniform and correct chemical content. There is no doubt that the general use of costly fluxed electrodes will seriously impede the extension of arc welding in many directions.

Direct current is generally preferred to alternating owing to the greater ease of striking and maintaining the arc, and also to the fact that the welder's hand soon becomes unsteady when alternating current is used. Good welders prefer direct current for overhead work. These factors have combined to make the use of direct current general in British practice. Good work has been accomplished with alternating current, but it is not recommended by makers of equipment.

The results of experiments made by members of the British Research Committee regarding the effect of voltage agree with those of Mr. Wagner. Some earlier outfits were constructed on the assumption that a minimum of 100 volts was essential; but it is now customary to employ 60 to 70 volts at the generator. The question of the effect of current is more debatable. From an analysis of the Wirt-Jones tests, the conclusion has been drawn that "the stronger the current the better the weld." This interpretation does not appear entirely justified, in view of the discrepancies presented by the results. For example, No. 31, with 150 amp. direct current, shows an ultimate tensile strength of 60,900 lb., while No. 9, also with 150 amp. direct current, shows only 38,100 lb. Until greater uniformity is obtained, it does not appear possible to generalize on this point. In some recent work, Dr. J. H. Paterson of the British Committee found that the effective maximum deposition was

obtained with 140 amp. when 20 gm. per min. were deposited. Beyond this strength, owing to splashing, the effective deposit diminished. Moreover, analysis showed that the minimum amount of oxygen entered the deposit at 140 amp. This point is urgently in need of settlement, with a view to guiding the design of equipment; and a series of tests, under identical conditions as far as procurable, should be conducted to decide the question. At least six welds, using identical plates and electrodes, should be made at each current strength to enable a correct decision to be arrived at.

With regard to the speed of arc welding, the maximum, according to Doctor Paterson's figures, would be 1200 gm. or about 2.6 lb. per hr., not allowing for intervals. Experiments confirming this would be useful, so as to settle this point definitely.

J. CHURCHWARD,* New York, N. Y. (written discussion†).—Mr. Hobart states that various kinds of electrodes are advocated, running from the cheapest fence wire at 10 c. per lb. up to the covered electrode costing five times that price. The importance of the electrode question has not been covered extensively enough. It is of the utmost importance to have an electrode that will be absolutely uniform in its welding qualities; that the deposited metal be as near the same chemical composition of the parent metal as it is possible to have in electric arc welding. If these things are accomplished, the price, within reasonable limits, should be disregarded.

It has been suggested that, when welding wire shows bad welding properties, it be dipped into whitewash to insure its uniformity of flow. The welder must take extra precautions to see that the slag introduced by this coating is puddled out of the weld. It is better to buy a good bare electrode than to coat a bad one, as with the latter there is always the danger of introducing slag into the weld.

The specifications for electrodes from the Welding Committee of the Emergency Fleet Corpn. are: Carbon not over 0.18 per cent., manganese not over 0.55 per cent., phosphorus and sulfur under 0.05 per cent., silicon under 0.08 per cent. The high limits are covered by patents issued to the Wilson Welder & Metals Co., in which an excess of manganese was allowed. The average manganese content for old welding wire was not over 0.35 per cent. If I understand these specifications correctly, they do not demand the rejection of any wire that would give satisfactory results even though not of this chemical composition.

Ductility of Welds Made with Covered and Bare Electrodes.—Relative to the ductility of welds made with covered and bare electrode welds, our data show that normally coated electrodes produce less ductile welds than bare electrodes due to the slag inclusions made with the former.

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* Metallurgist for Wilson Welder & Metals Co. † Received Mar. 29, 1919.

Effect of Voltage on Arc Welding.—Increased voltage, say from 40 volts up, enables the inexperienced operator to weld with greater ease on account of the length of the arc this voltage allows; but it sacrifices the strength of the weld. Experienced operators will not vary the length of the arc more than 3 or 4 volts, regardless of the open-circuit voltage of the machine. The voltage used across the arc by expert welders varies from 16 to 19 volts. With an inexperienced welder or learner, records show that a high open-circuit voltage has a tendency to make the operator careless and thereby cause bad welds.

Overhead Welding.—The quality of overhead arc welding depends entirely on the experience of the operator, the quality of the wire, and the characteristics of system used. With excellent wire, an expert operator, and a machine of the proper characteristics, overhead welding may be very easily as strong if not stronger than flat welding.

Different Lengths of Arc.—It is essential in metallic arc welding that the length of the arc be kept as short as possible. This introduces two variables, the quality of the wire and the expertness of the welder. An expert welder with a good wire will maintain a 16-volt arc; with a poor wire he will try to maintain the same length of arc with many interruptions due to the breaking of the arc. An inexperienced welder will maintain the length of arc that will enable him to deposit the metal with the least possible exertion. If the electrode is good the arc will be short. If the electrode is bad the arc will be long, if the design of the machine so allows. With a long arc, there is less heat per unit area in the parent metal and therefore less concentration of welding heat than with a short arc. If the voltage is so high that the operator is enabled to draw a long arc, although it may be easier for him, the weld will suffer. One would draw the conclusion from the above that with the personal element, the current and type of electrode constant, the quality of the weld depends on the length of the arc.

CHARLES H. KICKLIGHTER, Atlanta, Ga. (written discussion*).—The disk-depression method of electric welding was developed primarily as a simple and effective means for forming a highly efficient joint or seam between heavy commercial plates. After the preparation of the parts, by making flat-bottom depressions in the surfaces of the plates and introducing therein thin flat disks, the parts are brought into the proper position and subjected to pressure and heating electric current between the terminals of an electric welding machine. The intermediate bodies, or disks, are welded to both plates and are softened and forced into the depressions so as to permit the faces of the plates to come into immediate contact.

The diameter of the disk is slightly less than the diameter of the de-

* Received March 31, 1919.

pression, but the thickness of the disk is greater than the depth of the depression, so as to hold the plates out of initial contact. Flat disks are used as the best results are obtained by having the areas of initial contact substantially the same as the areas of the desired weld and applying the pressure normal to these surfaces. The surfaces of contact are thereby brought to a welding heat at all points at about the same instant and a thorough fusion of the metals at all points is obtained by the time the intermediate metal is softened sufficiently to be forced into conformity with the depression. A plurality of such welds, properly staggered, makes an efficient joint.

Description of Process.—Simple flat disks may be placed loose in flat-bottom depressions, as in Figs. 42 and 43. Here disks *3*, in depressions *4*, between plates *1* and *2* are subjected to pressure and the heating electric current between terminals *5* and *6* of an electric welding machine. The appearance of the disks, after the welding operation, is shown at *7*. In order to hold the disks at the centers of the depressions during the progress of the work, small register projections *8*, Figs. 44 and 45, on the disks fit snugly into register holes *9* in the bottoms of the depressions. Fig. 46 shows a special drill for cutting the depressions and register holes. The cutting edge *14* of the drill *13* cuts the depression *4* and the small auxiliary drill *15* cuts register hole *9*. The collar *16* limits the depth of the depression. A bank of, say, six, eight, or twelve such drills, operating simultaneously, will cut the depressions rapidly and economically.

The shifting of the disks may be prevented by sticking them to the bottoms of the depressions; by fusing them at one or more small projecting points of the contacting surfaces. Fig. 47 shows a disk *3* stuck in place by fusing a projecting central point *20* to the bottom of the depression *4*; Fig. 48 shows a disk *3* with two peened-over edges *10*, which are fused to the bottom of the depression *4*; and Fig. 49 shows a joint between two metal plates in which the disks *3* have been previously stuck in place at the peened-over edge points *10*. It will be noted that depressions may be cut in each plate near the edge and that the plates are brought into position with the depressions overlapping or falling beyond each other, so that, in the welded joint, the full stresses can never come on a plate where it has been weakened in the least by such depression.

Disk-depression Welding versus Riveting.—The advantages of the disk-depression method of electric welding over riveting are: a joint of greater efficiency and greater economy in production. A joint of very high efficiency is obtained, for the plates are not weakened by the cutting away of material in punching rivet holes. The small amount of metal removed in making the depressions is negligible and even this is restored by the disk material, which is fused into and completely fills the depressions. By cutting depressions in both plates near their edges and overlapping these, the full stress never comes on a plate where depressions

have been cut. The only reason why the joint may not be made as strong if not stronger than the solid plate is a slight tendency of the metal of the plates to crystallize when heated, but due to the small region heated and the low temperature attained in the greater part of this region, this weakening effect is very slight.

With proper means of bracing or mechanical framing, the seam can be completed for the present cost of rivets alone. The amount of heating current is relatively small and the heating is limited to the immediate position of the weld. As the welds can be made rapidly, the conduction of heat by the plates causes very little heating of other parts.

Disk-depression Welding versus Arc and Autogenous Welding.—The advantages of the disk-depression method of electric welding over arc or autogenous welding are: greater efficiency of joint; greater saving in time, labor, and electric power; and elimination of distortion troubles during heating and resulting stresses after cooling. Arc-welded or autogenous-welded joints are never as strong as the solid plate unless they are built up of considerably greater thickness than the plate. The great saving in time, labor, and power will be appreciated when one considers the length of time, labor, and amount of heat necessary to weld a seam between thick plates by the arc or autogenous method. With the disk-depression method, a few seconds per weld is sufficient.

Disk-depression Welding versus Spot Welding.—The advantage of the disk-depression method of electric welding over present methods of spot welding are: greater efficiency of joint; greater economy of welding current; smaller capacity and less cost of welding machines; practically no trouble because of distortion of work during operation due to heat; greater thicknesses of plates that can be welded; and simplicity and ease of preparing work prior to welding operation.

A high efficiency of the joint is insured by having clean surfaces free from scale. The operator is then certain that a perfect weld has been made when the disk becomes sufficiently fused to let the plates come into immediate contact. If the contacting surfaces are covered with scale, he cannot be absolutely sure of a perfect weld. The disks are of sufficient thickness to hold the plates out of immediate contact before the welding operation and thus serve to narrow the path of the welding current and concentrate its heating effect. The resistance to the welding current is small, thus requiring a minimum amount of electric power and a welding machine of relatively small capacity. With clean surfaces, the mechanical pressure does not have to be so great as where there is scale or the plates have to be forced into contact about an intermediate body with no depression in connection therewith. For this reason, a welding machine of lighter construction and smaller cost will perform the work.

The greater part of the resistance to the welding current is at the surfaces where the disk contacts with the plates. Here the heat is concen-

trated and in a few seconds it becomes so intense as to fuse the thin disk to both plates before the surrounding parts of the plates become heated by the conduction of heat. The seam does not become nearly so hot as in other methods of welding, the distortion of the work due to expansion is not nearly so great, and the stresses set up in the plates due to cooling and unequal contraction are small. As most of the heat is developed and concentrated at the contacting surfaces of the disk, plates of considerable thickness may be easily welded. As to the thickness of the material, there does not appear to be other drawback than the matter of forcing the plates through a very small distance into close contact. As the region of intense heat is so small and the weld is performed before the heat has spread much by conduction, the region affected by any tendency to crystallize is very limited and the strength of the joint is not appreciably affected thereby.

As the depressions are very shallow and do not have to be accurately spaced, they may be rapidly cut by a special cutter or drill with an attachment limiting the depth of cut. A bank of six, eight, ten, or twelve drills will permit the cutting of a corresponding number of depressions in the time otherwise required to cut one. It is preferable to cut the depressions before the plates are bent into shape.

The slight increase in cost and the slight loss of simplicity occasioned by cutting the depressions is far outweighed by such considerations as the insuring of a perfect weld, a highly efficient joint, a large saving in electric power, the use of lighter and less expensive welding machines, rapidity of the welding operation, freedom from heat-distortion and stresses set up by the subsequent cooling and contraction, less crystallization of the metal due to heating, etc.

Theory of Method.—Fig. 50 illustrates the contracted path of the heating electric current as it passes between the broad flat-end terminals *5* and *6* through plates *1* and *2*. Most of the resistance to this current is along the surfaces of contact *aa* and *bb*, and most of the heat is developed in the disk *3* and in plates *1* and *2* in immediate contact therewith. Fig. 51 shows how it is often attempted to concentrate the pressure by means of conical terminals *5* and *6* in order to obtain a weld between plates *1* and *2*. Due to the rigidity of the thick plates, this pressure is distributed over a wide range between the plates *1* and *2*. The heating current also spreads considerably as *cc* where plates *1* and *2* are in contact. The path of the current is contracted most at *dd* and *ee*, causing considerable loss of heat, marring of the surfaces of plates *1* and *2*, and damage to terminals *5* and *6*.

Fig. 52 illustrates a disk *3* placed between two plates *1* and *2* and welded thereto by current passing between terminals *5* and *6*; if it is now attempted to force the plates into immediate contact, as in Fig. 53, the disk flattens and spreads out as at *7*. The area *ff* may be welded if the

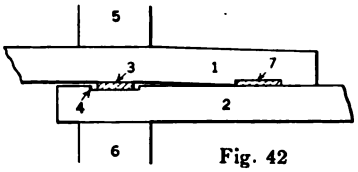

Fig. 42

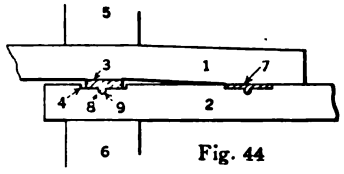

Fig. 44

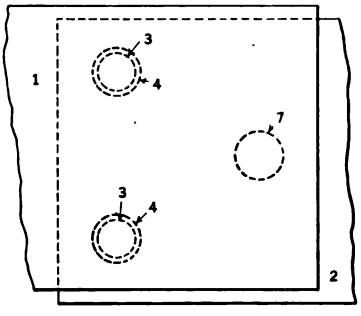

Fig. 43

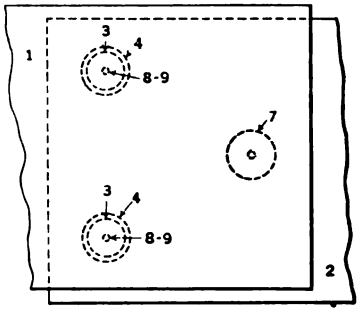

Fig. 45

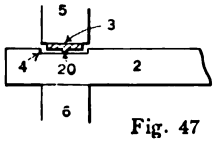

Fig. 47

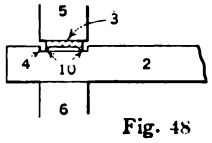

Fig. 48

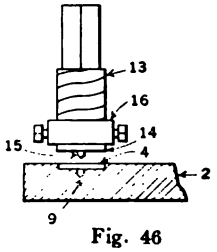

Fig. 46

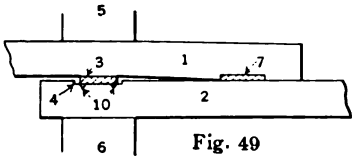

Fig. 49

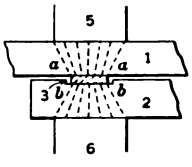

Fig. 50

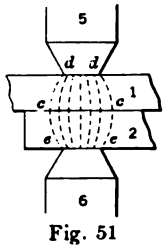

Fig. 51

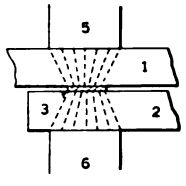

Fig. 52

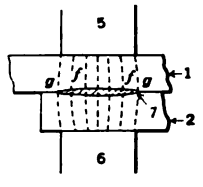

Fig. 53

DISK-DEPRESSION WELDING.

surfaces are clean, but the areas *fg* where the molten metal has been squeezed out against the cold plate surfaces is not welded; there is also some difficulty in forcing the plates into immediate contact around the weld. There is considerable widening of the path *gg* of the current between the plates; this causes considerable heat in the joint, resulting in heat-distortion and subsequent cooling stresses. The region of partial crystallization is greatly enlarged.

Tests.—In order to determine the efficiency of the disk-depression method of electric welding, commercial bars $\frac{3}{8}$ in. thick and 1 in. wide were lapped and welded, some in two places and others in three. Depressions were cut in one of the bars $\frac{7}{16}$ in. in diameter and $\frac{3}{64}$ in. deep. Disks $\frac{1}{16}$ in. thick cut from a $\frac{3}{8}$-in. commercial round bar were placed between blunt flat welding terminals of an ordinary spot-welding machine. The capacity of this machine in ordinary spot-welding work was $\frac{1}{8}$-in. plates or less, but we were enabled to get splendid welds on stock $\frac{3}{8}$ in. and $\frac{1}{2}$ in. thick by the disk-depression method. Some samples were grasped in the jaws of a vise and hammered over until they tore apart, revealing a perfect weld and a fibrous structure. Others were tested in a tensile testing machine until they broke. Bars with two welds broke at 13,090 lb. and those with three welds at 19,405 lb. The solid stock broke at 19,275 lb. and had an elastic limit of 13,125 lb.

ZAY JEFFRIES,* Cleveland, Ohio (written discussion†).—The following is quoted from a letter, dated Sept. 9, 1918, to Mr. Hobart.

"In reference to the improvement of electric welds by heat treatment, I have observed that the steels higher in carbon are helped more by heat treatment than the low-carbon steels. High-carbon steels are often the cause of a martensitic structure in the weld and, of course, the higher the carbon the more brittle will the martensite be, and the more will ductility be restored by heating above the critical temperature and cooling slowly. I have observed well-defined martensite in welds made with piano wire and with a nickel steel containing about 0.2 per cent. carbon and 3.5 per cent. nickel. As the carbon decreases, the martensitic needles become farther apart and the steel becomes more ductile.

"In addition to the martensitic needles, there are other needles to which Mr. Ruder has referred, which he seems to have proved to be nitride of iron, or at least the presence of the needles if they are not iron nitride seems to be due to the presence of nitrogen. We must not conclude prematurely, however, that the needles themselves are composed of nitride of iron because Timofeeff shows the development of the same microstructure by straining commercially pure zinc at room temperature. There is no sign of this structure in the annealed zinc so that the needles must be a direct result of the straining. At the same time, Howe has

* Director of Research, Aluminum Castings Co. † Received Mar. 24, 1919.

produced needles of a similar nature in manganese steel and there is no reason to suppose that the composition of the needles is any different from that of the remaining mass of the metal.

"But it has not been demonstrated that welds containing needles are always brittle and that welds not containing these needles are always ductile. In fact, I have examined many welds under the microscope that did not show the needles, but which were brittle.

"Recently J. C. Lincoln, of the Lincoln Electric Co., Cleveland, Ohio, gave me four samples of electrically welded steel for microscopic examination. These were made with a modified carbon-arc process. Two steel plates $\frac{3}{8}$ in. thick were placed edge to edge on a flat surface and the first side welded, then the plates were turned over and the other side welded. Bending tests were made in such a manner that the two plates bent parallel with the weld. The bending was stopped when the first crack appeared in the weld. Weld No. 1 was made in an atmosphere of carbon dioxide and a crack appeared after bending 90°. Weld No. 2 was made in an atmosphere consisting of about 90 per cent. nitrogen and 10 per cent. carbon dioxide and bent 60° before cracking. Weld No. 3 was made under flux and it bent 50° before cracking. Weld No. 4 was made in the air and a wide crack opened up after bending 45°. Welds Nos. 1 and 2 are unusually ductile for electric welds and they also contain the needles. Weld No. 3 contains the least number of needles, but it is not the most ductile. Weld No. 4 contains needles and it is the most brittle of the four. This weld, by the way, is an unusually good air weld.

"Having found, therefore, that electric welds containing needles may be ductile and electric welds containing no needles may be brittle, it seems evident that the needles are not the main cause of brittle welds. In this connection I should like to present J. C. Lincoln's opinion, namely, that oxygen is the "embrittling element" in electric welds. He has reached this conclusion after many and varied tests. The same idea is strongly suggested by the most interesting results reported by Mr. Wagner with magnesium and boron suboxide. Both of these would be oxygen removers. That the magnesium did remove oxygen seems apparent from Mr. Wagner's remark that the liquid metal in the weld "boiled" when the magnesium came in contact with it.

"Iron oxide can be seen in the welds under the microscope in the form of globules. In a microscopic examination we cannot hope to see particles of iron oxide containing less than about a million molecules. Owing to the nature of the transfer of metal in the making of an electric weld and the rapidity of solidification, it would not be surprising if much oxygen was present in the form of iron oxide in particles too small to be seen under the microscope. So far as the microscopic examination is concerned, these particles might actually be in solution. It is, therefore, possible

that the "embrittling element," or at least the portion of the element which is the most harmful, is present in such small particles that it cannot be seen under the microscope.

"The weld made by Mr. Wagner with Armco iron weld rod coated with magnesium powder is the most ductile electric weld I have seen.

"It seems to me that this lead is so promising that it should be followed with the utmost speed. If it works out commercially, as well as in the sample submitted, it is my opinion that the matter of the quality of electric welds will be quite satisfactory. If the weld is ductile, it need not have as high tensile strength as when it is brittle."

It seems probable that Mr. Lincoln's experiments in no case entirely excluded nitrogen, but it is probable that there was a smaller percentage of oxygen in the atmosphere surrounding the welds where gases other than air were used than in the weld made in the open air. The weld made in an atmosphere of 90 per cent. nitrogen should have less oxide between the grains than that made in air, but there seems to be no reason why it should have less nitride present than the one made in air. The weld made in the 90-per cent. nitrogen atmosphere, however, was more ductile than that made in air, which would seem to bear out Mr. Lincoln's conclusion that oxygen is the "embrittling element."

It may be definitely concluded that nitrogen is the element that embrittles the grains. If the grains contained less nitride and hence had lower elastic limits, the grain-boundary strength might remain unchanged (that is the same as now observed) and the weld would show considerable increase in ductility on tensile test, without necessarily decreasing in strength. It would, therefore, seem advisable to keep the nitrogen content low in electric-steel fusion welds.

It also seems that the grain-boundary weakness is due to either oxygen or nitrogen or both. It remains to be decided how much each of these elements contributes to the weakness of the grain boundaries and to eliminate one or both from the welding atmosphere if sufficiently beneficial.

A. S. KINSEY,* New York, N. Y. (written discussion†).—On page 588 the author states that "experience in arc welding has gradually been acquired and it controverts the correctness of the view" of a gas-welding publication quoted which shows the advantages of the oxyacetylene over the electric-arc process. I presume the author refers only to the statement of the limits of the arc weld, and that even though there may be an absence of acceptable test data on gas welds on which to base his judgment, he will agree that oxyacetylene welds have, when carefully made, proved highly satisfactory to the best engineering practice, both in ductility and tensile efficiency. Also, there surely can be

* Advisory Engineer, Service Dept., Air Reduction Sales Co.
† Received Apr. 5, 1919.

no doubt in the mind of the author as to the accuracy of the statement of the quoted article as to the value and adaptability of the oxyacetylene torch for annealing a steel weld. The reference to the absence of satisfactory laboratory tests of oxyacetylene welds need bear no undue significance, as the real situation has been that the oxyacetylene welds were so much stronger and more economical than the forged welds that tests of the physical value of gas welds were never necessary. They have stood the test of actual practice. On the other hand, is it not true that in the development of the electric-arc weld for use on steel plates for ships, it has been necessary to know exactly what are the physical limits of the arc weld in order to determine the risk connected with its use on ship hulls? This is what caused laboratory tests to be made of arc welds.

H. G. KNOX,* Norfolk, Va. (written discussion†).—There is one subject that Mr. Hobart seems to have inadvertently omitted from his very valuable paper, and that is corrosion. The shipbuilder, perhaps more than anyone else, has constantly to guard against it, our chief reliance being paint and galvanizing, where practicable.

The corrodibility of a welded joint in salt water must be determined before deciding whether this type of construction is to supersede riveting. At the present time, this question cannot be answered, although considerable work has been started in an attempt to arrive at a conclusion. The British Admiralty and several British shipbuilders have studied it extensively. Their experiments and conclusions have been very carefully followed and reported in this country by Naval Constructor Goodall of the British Navy. To date, these show that no undue fear need be felt from corrosion. At the Norfolk Navy Yard a series of plates welded with various electrodes were submerged in dilute sulfuric acid, such as is used in cleaning prior to galvanizing. While this test may not be conclusive in any quantitative way, it does show a marked difference in the corrodibility of welds made with different electrodes under different conditions. To draw any conclusions from this accelerated test would be, however, ill advised.

In riveted joints, the exposed heads and points of the rivets always show the greatest tendency to corrode. This is due to two reasons: first, as they are raised above the surrounding surface the paint is more susceptible to abrasion; and, second, electrolytic action is set up between rivet and plate. These same reasons lead us to expect a tendency of the welded joint to corrode.

Various methods have been suggested for decreasing the probable inherent corrodibility of the weld, in addition to the customary coat of paint. These generally embody the introduction of alloys, principally

* Commander, U. S. N., Construction Corps. † Received Apr. 12, 1919.

copper, into the electrode material. Plating the electrode with copper is not effective in that the deposited copper segregates in the weld and actually increases the electrolytic action. As far as is known, no extensive test on the so-called copper-bearing alloy steels, which might be made up into electrodes, has been carried out, but when obtainable, electrodes of this alloy will be subjected to service tests.

About 9 mo. ago, we started attaching small plates, welded with different manufacturers' electrodes, to the under-water hulls of steel ships placed in dry dock. Up to the present time none of these ships have redocked in the Norfolk Yard, so no results from this source are available. We have also used several makes of electrodes in the construction of an arc-welded battle towing target, which will soon be launched.

FIG. 54.—ARC-WELDED BATTLE TOWING TARGET UNDER CONSTRUCTION AT NORFOLK NAVY YARD.

It can readily be seen from the foregoing that little definite information on the subject of corrosion is available, but for the time being, at least, I do not believe any special precautions need be taken in the protection of welded joints beyond that employed for the surrounding plates.

JOHN MARTIN,* New York, N. Y. (written discussion†).—It was the writer's privilege to represent the American Bureau of Shipping on the Welding Research Sub-committee over which the author of the paper under discussion presided with notable tact, consideration and distinction. The foreword that the paper deals principally with investigations under-

* American Bureau of Shipping. † Received Apr. 29, 1919.

taken by the Welding Research Sub-committee relieves the author of any save editorial responsibility for certain irreconcilable statements, some of which might be expected in the experimental research data of any new industry and especially in one like metallic-arc welding, which introduces so many variable factors and conditions essential to the process that a long acquaintance with the technic and data records is required before one can detect and discard an irreconcilable result.

Page 581 states that a Norway bare electrode wire contains 0.007 per cent. sulfur and that the deposited metal of weld contains 0.072 per cent. sulfur. As the ½-in. ship plate used for testing electrodes held 0.032 per cent. sulfur, where did increased sulfur come from? The metallic arc is a reducing agent and no flux was used. This illustration is used to show how carefully all test reports should be scrutinized and compared before insertion in finished records and in no sense to detract from the value of the author's compilation.

Composition of Ship-plate Steel.—Where ship-hull plates are united metallurgically (welded) instead of physically (riveted), the chemical composition of the plate will become as important as the physical performance of its test specimen prior to installation. A certain amount of slag (reduction of foreign matter, carbon, manganese, phosphorus, sulfur, and silicon) is driven off in all fusions whatsoever the welding process used (smallest, however, in spot welding) and the chemical composition of the ship plate must be such that, after welding, stresses will be transmitted through the joint as efficiently as through the plate, that is, the joint shall be 100 per cent. efficient the same as many riveted joints. This condition is impossible of accomplishment by arc welding from our present knowledge of the art as the deposited metal in the weld is always pure iron whatever the chemical composition of the electrode may be. In the spot weld, although the writer has no knowledge of their chemical composition, the "spots" more nearly approach the original composition of plate as the slag driven off (a measure of the chemical change within metal) is inconsiderable compared with that of arc welds.

Fusion Welding and Spot Welding.—The author distinguishes between welds as fusion welds and spot welds. Spot welding is fusion welding and to differentiate by grouping under fusion is apt to create a false picture in the uninitiated reader's mind. It is suggested, starting with the premise that a weld with heat means fusion, to classify welds processively as spot, gas, carbon-arc, metallic-arc welds in which the physical difference is that of continuity of weld.

Metallic-arc Welding.—Inasmuch as overhead vertical and horizontal welds are equally well performed with the metallic arc, it may be assumed that the flow (or bombardment of metallic molecules) of metal from the electrode to the joint follows Newton's law of attraction of matter (directly

as the mass and inversely as the square of the distance). Molecular cohesion of the electrode molecules, lessened by the heat of the arc, is overcome by the attraction of like mass in the plates; the result is the passage of electrode molecules to the plate, whence adhesion ensues, which becomes cohesion upon cooling contraction and consequent crystallization of deposited material.

Spot Welding.—The author's description of spot welding is satisfactory save that no initial voltage is specified. In one test performed in the writer's presence 32,000 amp. with a voltage of four successfully spot-welded pairs of ⅝-in. to 2½-in. (15 to 63 mm.) ship steel plates together, the diameter of the spots being 1 in., the pitch 3 in., and the hydraulic pressure 7½ tons, the time of the current was 15 sec. and 15 ft. was welded in 30 min. The spots sheared at 56,000 lb. or partly pulled and sheared at 60,000 lb. Water- and oil-tight joints resulted without calking. For plates of the same pitch calking or arc fillet welds would be required for water-tight joints; though by lessening the pitch and decreasing the diameters of the spots, practically the same water-tight results could be had.

Summarization of Points for Discussion.—From the fact that the author has found 34 points, the discussion of which he believes may advance welding technic but only one of which concerns spot welding, leads one to infer that much is to be learned about arc and little about spot welding. The writer believes that the greatest obstacle to the adaptation of spot welding to hull construction is the difficulty of manipulation of spot-welding apparatus in and about the hulls of ships, and that is but a mechanical detail to be worked out. If a year ago it had been realized that there were 33 variables in the arc-welding and but three in the spot-welding problem, the latter would have determined perhaps the more fruitful field for experimental investigation in seeking a satisfactory weld for ship construction.

The writer differs with the author's contentions that the best basis to express the speed of arc welding is in pounds of metal deposited per hour. Feet of joint per hour welded would better assist the ship builder in his estimates and cost sheets and certainly would afford a simpler basis for comparison between welding methods. The writer heartily indorses the author's belief that spot welding has a great future as applied to shipbuilding but does not agree with the statement that spot welding affords a method of preliminarily joining the hull plates after which the required strength is furnished by arc welding. The writer would have expressed the thought thus: spot welding affords a method of joining the hull plates, after which the required water tightness may be secured by calking or calking arc welding.

The author points to the future application of spot welding to hull construction when he finds that the strength of the spot-welded joint,

with calking arc weld, is 50,000 lb.; of the arc-welded joint, with calking arc weld, is 37,000 lb.; and of the riveted welded joint, with calking arc weld, is 13,000 lb. In other words, the joints are as 4 : 3 : 1, respectively. That is, the spot-welded joint with calking arc fillet is one and one-third times as strong as the arc-welded joint and four times as strong as the riveted joint.

Conclusion.—The writer agrees with the author in his conclusion that welding should be applied in the general fields where the hazard to life would be minimum before applying it to constructions in which joint failures would jeopardize human lives. But he deprecates the author's under-rating the stresses in ships as being less severe on welded joints than those in high-pressure containers. The stresses in ships are dynamic tensile and compressive, torsional, alternating, and an admixture of all these, whereas those in a high-pressure container, even in marine boilers, are largely static and it is the alternations of stresses that fatigue ship members and require the ever watchful eye of the classification society to conserve property and preserve life.

Cooling Properties of Technical Quenching Liquids

BY N. B. PILLING,* M. S., AND T. D. LYNCH,† C. E., PITTSBURGH, PA.

(Chicago Meeting, September, 1919)

THE development of a proper treatment for shells in connection with war contracts has brought to our attention the fact that the temperature of the liquid bath in which steel is quenched has a decided influence on the speed of cooling in the case of some liquids and very little in the case of others. Widely different physical results have been obtained by the use of water spray, water immersion, and oil immersion and further experiment has shown that the quenching effect of water used as a spray or an immersion varies greatly with the temperature of the water.

The present experiments were undertaken with the idea of obtaining information of a quantitative nature regarding the quenching properties of a number of liquids used commercially in hardening steel. The general features of such cooling media are common knowledge and in the literature may be found records of numerous attempts to evaluate the cooling power of a liquid when used as a quenching bath. Much, if not most, of the data given on this subject is in the form of a simple statement of the time required for a certain heated mass to cool between two arbitrary temperatures when immersed in the liquid, quite regardless of the path of the temperature change. Our own purpose has been somewhat different. We have attempted to isolate that property of these liquids having to do with their ability to absorb heat rapidly, as distinct from the behavior of a considerable mass when exposed to different thermal environments. More specifically, from the data obtained through autographically recorded cooling curves of a small standard mass quenched under various conditions, we have tried to analyze the mechanism of heat absorption by the several liquids, measure the changes in its rate as a function of the temperature of the surface of transference, and ascertain the influence of the temperature of the liquid itself upon both.

Without attempting an exhaustive citation of the literature, mention may be made of some of the more important contributions, notably the pioneer work of Le Chatelier,[1] followed by the extensive investigations

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[1] Le Chatelier: Etudes sur la Trempe de l'Acier. *Rev. Met.* (1904) **1**, 473.

of Benedicks.[2] Although working with samples of small mass and with sensitive instruments, the data presented by them is open to the objection, serious from the point of view of the thermal action of the quenching medium, that steel is quite unsuited for use as a heat reservoir due to the uncontrollable liberation of heat at the partly suppressed transformation masking the temperature changes due to the extraction of heat by the cooling liquid. Matthews and Stagg[3] have contributed valuable information on the actual cooling of steel pieces, large enough to time with a watch, in certain liquids and have measured the effect of increasing

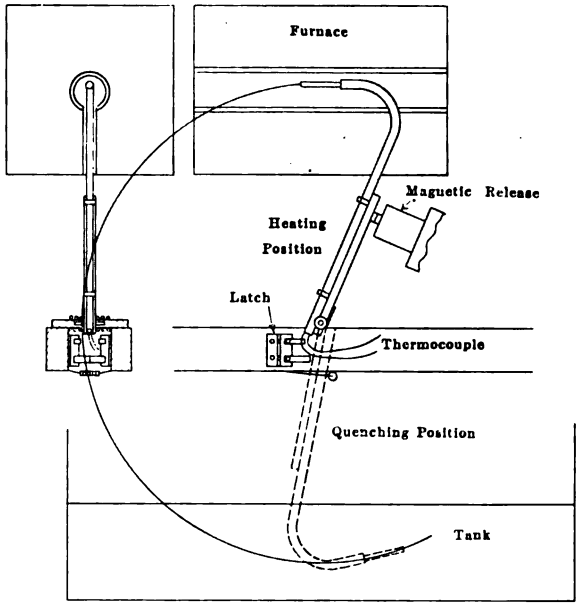


FIG. 1.—QUENCHING APPARATUS.

the bath temperature on the time required to cool from 650° to 370° C. From autographic cooling records of small cylinders of several metals quenched in (presumably cold) water, Garvin and Portevin[4] concluded that it was not feasible to represent the cooling curves mathematically. Fry[5] and Law[6] have given cooling records of steel forgings weighing ½ ton and upward in a number of liquids.

[2] Benedicks: Cooling Power of Liquids, Quenching Velocities, etc. *Jnl.* Iron and Steel Inst. (1908) **77**, 153.

[3] Matthews and Stagg: Factors in Hardening Tool Steel. *Trans.* Am. Soc. Mech. Engrs. (1914) **36**, 845.

[4] Garvin and Portevin: Etude Expérimentale du Refroidissement de Divers Métaux par Immersion dans l'Eau. *Compt. Rend.* (1917) **164**, 783.

[5] Fry: Notes on Some Quenching Experiments. *Jnl.* Iron and Steel Inst. (1917) **95**, 119.

[6] Law: Effect of Mass on Heat Treatment. *Jnl.* Iron and Steel Inst. (1918) **97**, 388.

In this investigation a systematic examination has been aimed at; and in this paper the results regarding seven liquids will be given—three oils, three aqueous solutions, and sulfuric acid. The general characteristics of these three types of liquid are briefly: The quenching value of water and aqueous solutions is high at low temperatures, is quite sensitive to changes in temperature, and falls off sharply to a low value

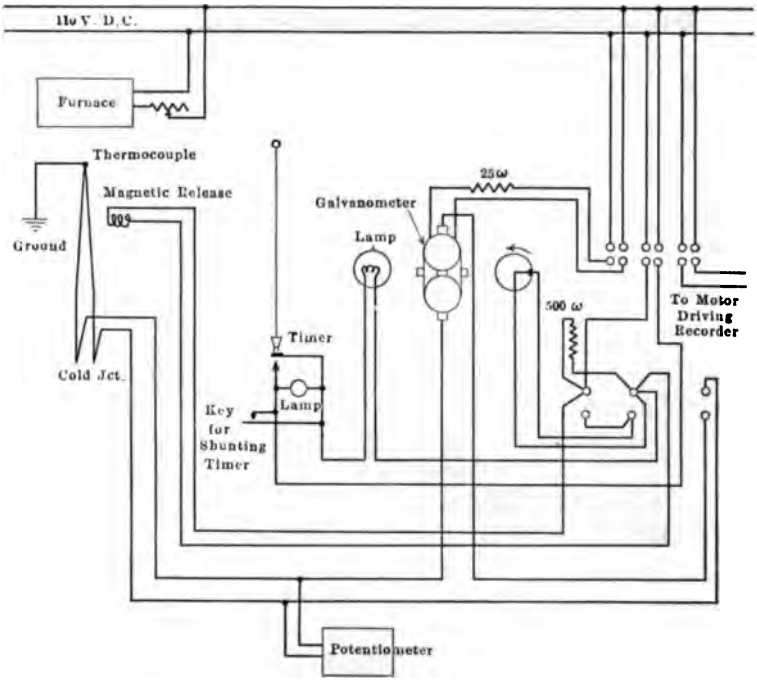


FIG. 2.—ELECTRICAL CONTROL OF QUENCHING APPARATUS.

with a moderate rise in temperature; oils with a moderate quenching value maintain their characteristics with but slight change to the maximum temperature reached; sulfuric acid partakes of the characteristics of both water and oil—high initial quenching value and stability at high temperatures.

OUTLINE OF METHOD AND DESCRIPTION OF APPARATUS

The method used in obtaining these measurements involved the accurate recording of the fleeting temperatures within a metal cylinder heated to a fixed initial temperature and rapidly transferred to a bath of the quenching liquid. This requires: (1) A temperature-indicating device of minimum thermal capacity, capable of faithful response to rapid temperature changes. (2) A temperature-recording instrument

sufficiently sensitive to respond to the fluctuations of (1). (3) A quick and uniform means of transferring the metal cylinder from furnace to bath.

In devising an instrument to fill these requirements free use has been made of the construction described by Benedicks,[7] and a diagram of the apparatus built is shown in Fig. 1. The key structure around which the rest of the apparatus was built is an iron tube $\frac{5}{16}$ in. (7.9 mm.) in diameter and about 12 in. long, clamped to a stiffening support of rectangular brass rod. At its lower end it was pivoted, swinging in a vertical plane; another tube bent into a right angle with a 2-in. radius and made of a non-scaling, heat-resisting alloy was screwed on to its upper end, forming

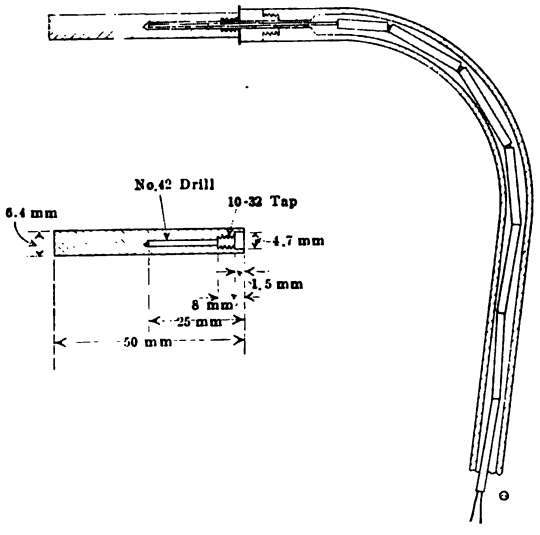


FIG. 3.—STANDARD CYLINDER AND THERMOCOUPLE.

an L-shaped structure. This bent tube served as a support for the cylinder whose cooling rate was to be measured and through which a thermocouple was introduced to the center of the cylinder.

A small box-shaped furnace with a horizontal alundum tube was suspended over, and in the plane of rotation of, the pivoted holder. A longitudinal slot in the lower side permitted the end bearing the cylinder to be rotated up into the heated portion of the furnace. An electromagnet held the holder in this position against the force of a coiled spring operating to rotate it out of the furnace. Below the furnace, on the opposite side of the pivot and in the path of the rotating cylinder was the tank holding the quenching liquid.

The thermocouple was made from platinum-platinrhodium wires 0.2 mm. in diameter and was threaded through lengths of two-hole

[7] Benedicks: *loc. cit.*

usalite capillary tubing, insulating the wires from each other and from the iron supporting tube. It is essential to use wires of as small diameter as possible, to reduce the error caused by the storage of heat in them and consequent false elevation of temperature of the junction during the rapid cooling. The last few inches of the couple were insulated with parallel lengths of one-hole usalite in place of the two-hole, reducing the insulating mass quite considerably. The bead at the tip must be in good contact with the cylinder, both thermally and electrically. The latter, as it leaves the furnace (temperature 830° C.), has a positive electrostatic charge which will slowly leak off through the thermocouple circuit; to avoid this it is necessary that the hot cylinder be permanently earthed. It is also required that the tip be pressed into the cylinder firmly enough not to be displaced in the least by the sudden jar caused by the latch; fault in this respect leads to induced currents in the couple circuit. Defective insulation of the couple wires from the supporting tube results in erratic parasitic thermoelectric currents causing irregularities of large magnitude.

The recording instrument used was a string galvanometer, equipped with a 0.7 mil silver wire as the moving element. This had sufficient sensitivity and an extremely short period. To it was attached a camera and rotating cylinder bearing an oscillograph film, upon which the image of the silver wire was projected (magnification 120×) moving at right angles to the direction of movement of the film. The deflection of the image recorded the temperature, and the time intervals were impressed upon the photographic records by suitable connections between a seconds beating pendulum and the electric lamp illuminating the galvanometer whereby it was winked at 1-sec. intervals. The instrument was frequently calibrated against a potentiometer during the course of the investigation and was found to maintain constancy quite well if care was taken not to handle the tension regulators.

A standard cylinder made from nickel + 5 per cent. silicon was used; it is shown in Fig. 3 attached to the holder. The greater ease of manipulation associated with the slower cooling rates of large masses is seriously offset by the temperature lag from the surface, where the heat exchange is operating, to the center where the temperature measurement is made, and a compromise between the diameter of the cylinder and the sensitivity of the recording apparatus must be made. The cylinder used was 50 mm. long, 6.4 mm. in diameter, and weighed 11.6 gm. A small axial hole drilled to its center permitted the insertion of the thermocouple, which pressed firmly against the metal at the base when screwed on to the supporting tube. A water-tight joint between the cylinder and the holder is essential, of course, and was obtained by the use of a small gasket of soft copper stamped from 0.010 in. (0.25 mm.) sheet. Steel is absolutely unfitted for a heat reservoir as the quantity

of heat liberated at the transformation temperature is both great and variable enough to mask the temperature changes due to the abstraction of heat by the cooling liquid alone. Nickel alloyed with 5 per cent. silicon was found to be quite satisfactory in this respect, as the silicon eliminates the nickel transformation and confers the additional desirable property of resistance to oxidation at high temperatures. The oxidation after long use was confined to a thin and adherent film, almost varnish-like in its tenuity and tenacity, and produced no change in surface con-

FIG. 4.—QUENCHING APPARATUS.

ditions. This permitted the repeated use of the same cylinder and avoided uncertainties due to possible small changes in mass and position of the thermoelement. The thermal conductivity and specific heat of this material are probably not greatly different from those of nickel, which is similar to steel in these respects.

The quenching liquid was contained in a galvanized iron rectangular tank placed below the furnace in such a position that the cylinder in rotating out of the furnace would be submerged in it. The volume of the liquid used was maintained at 17.5 l., corresponding to a depth of 12.5 cm. The operation of quenching the standard cylinder from above 800° C. raised the bath temperature but 0.5° C. Three electric heaters (toaster

stoves) beneath the tank heated the liquid to the desired temperature, measured with a mercury thermometer after thorough stirring. A latch engaged the rapidly moving iron tube bearing the cylinder after submersion and held it stationary in the liquid while cooling took place. About 0.5 sec. was taken in transferring the cylinder from the furnace to the latched position in the quenching bath, and temperature records during this interval showed no observable change in temperature of the

FIG. 5.—RECORDING APPARATUS.

cylinder. The temperature of the cylinder previous to quenching was 830° C. in all cases.

In operation, the liquid bath was first brought to the desired temperature and the cylinder placed in the furnace, being held in position by the electromagnet. Its temperature was followed during heating with a Leeds & Northrup potentiometer indicator. When the quenching temperature was reached, the photographic film was started turning at a suitable speed and the camera shutter opened. The magnet holding the cylinder and the 100-watt N. F. lamp illuminating the galvanometer were in series and the resistance of the magnet coils was enough to dim the lamp to a very dull glow. An electric contact was so placed in the driving mechanism that when the beginning of the recording film passed the camera window the magnet was shunted and the full voltage applied

to the lamp. Simultaneously, then, the cylinder moved out of the furnace into the quenching bath, impelled by the spring, and the lamp flashed up starting the photographic record. A switch was then closed shunting the magnet permanently. The time recorder was made of a seconds beating pendulum with a brass bob, to the nose of which was fastened a small U-shaped permanent magnet (made from hardened clock spring). The swinging pendulum passed over a light iron armature, which was raised by the magnet inserting a resistance in series with the galvanometer lamp and dimming it momentarily. A diagram showing the electrical scheme for the control of the apparatus is given in Fig. 2. Views of the quenching apparatus and recorder are shown in Figs. 4 and 5.

Composition and Properties of Liquids Examined

Data are given on the results of the systematic examination of the following liquids, all used technically as quenching media: Water, brine, soap solution, oil No. 1, oil No. 2, oil No. 3, sulfuric acid.

Water.—This was taken directly from the tap at East Pittsburgh. An average analysis is: 150 parts per million of solids.

Brine.—Made by dissolving common salt in tap water. Its analysis is: total solids, 24.00 per cent.; NaCl, 23.99 per cent.; density, 1.20.

Soap Solution.—About 2 per cent. of Ivory soap was dissolved in hot water. Stronger concentrations than this solidified when cold. This solution was a thick jelly at 20° C. but fluid at 30°. Its analysis is: total solids, 1.85 per cent.; true soap, 1.47 per cent.

Oils.—Three mineral oils were used having the following properties:

	Oil No. 1	Oil No. 2	Oil No. 3
Flash point, degrees C	188	204	305
Fire point, degrees C	215	241	341
Boiling point, degrees C	378	402	460
Density	0.882	0.912	0.916

The boiling point was taken as the temperature at which the evolution of vapor became voluminous.

Sulfuric Acid.—Concentrated acid was used. Its analysis is: H_2SO_4, 89.5 per cent.; H_2O, 10 per cent.; total solids, 0.34 per cent.; density, 1.810.

Quenching Action of a Liquid

The autographic cooling curves are not suitable for direct comparisons on account of the unequal values of the time spacings, necessitated by regulating the speed to receive the record on one length of film, and on account of the distorted temperature-deflection characteristics of the galvanometer. For purposes of visual comparison, selected curves illustrating the behavior of each liquid have been replotted to uniform

temperature and time scales. Strict comparisons may be made between these cooling curves as the sole variable was the temperature of the quenching bath. A study of these time-temperature cooling curves has shown that certain generalizations regarding the mechanism of heat abstraction

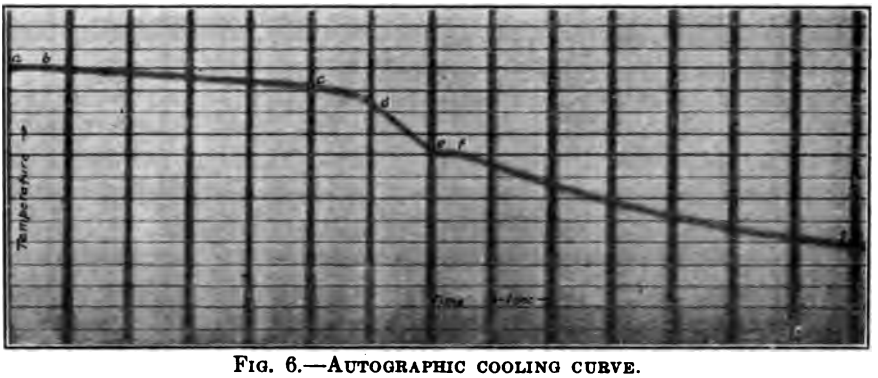


FIG. 6.—AUTOGRAPHIC COOLING CURVE.

by a liquid from a heated solid may be made, and from an understanding of these principles a more rational interpretation of the results to be given may be expected.

Fig. 6 is the autographic cooling record of the standard cylinder when quenched in tap water at a temperature of 58° C. from an initial temperature of 830° C., and Fig. 7 is this same curve transferred to a uniform

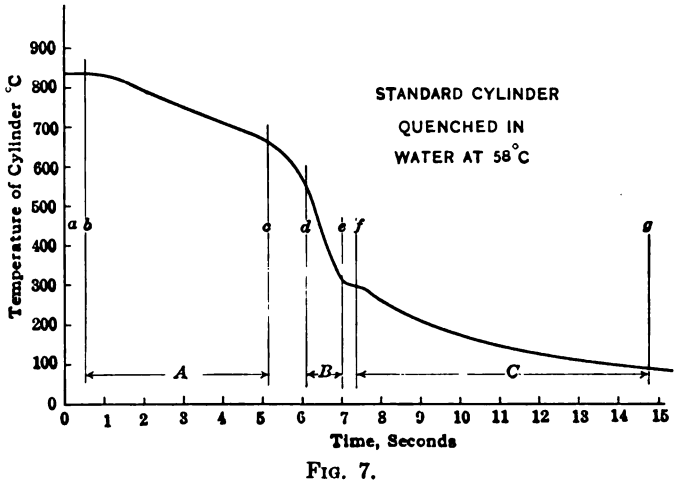


FIG. 7.

temperature scale. The record starts at *a*, with the cylinder leaving the furnace; at *b* it enters the quenching bath and at this instant the water in immediate contact with the surface of the cylinder is quickly raised to the boiling point and vaporized, forming an envelope of steam around the cylinder. The cooler layers of water strive to condense this steam,

while heat flowing from the hot cylinder tends to superheat it. The survival of this steam jacket depends on the relative rates at which heat is delivered through the jacket to the boundary surface between steam and water, and is removed from it by the adjacent cooler layers of water. In this case the former is the greater and the vapor jacket, with its relatively poor thermal conductivity, persists. The branch of the curve between *b* and *c* represents the sojourn of the cylinder in this atmosphere of steam. At *c* the two rates become equal, reverse in magnitude, and at *d* the vapor jacket disappears. There is now liquid contact between the heated surface and the water, and as the temperature of the metal surface is far above the boiling point of water, vigorous boiling ensues, each minute bubble of steam formed being whirled away by the forceful local convection currents and immediately condensed. From *d*, where a steady loss of heat by this means is attained, to *e* the liquid shows a great capacity for absorbing heat quickly and the temperature fall is very rapid. As a result of this vigorous campaign of vaporization and condensation, a concentric layer of water of appreciable thickness at the boiling temperature accumulates and the quick cooling terminates at *e*. Further absorption of heat by this concentric layer tends toward the re-establishment of the vapor jacket and a virtual halt in the cooling results. About 0.3 sec. is now consumed in cooling this layer before the cooling of the cylinder is resumed at *f*, and from *f* to *g* the heat absorption, relatively slow, is by thermal conduction into water below the boiling point.

To summarize, the rather complicated mechanism of heat transfer during quenching may be divided into three stages, arranged in a descending temperature sequence:

Stage	Character	Designation
Cooling in vapor	Slow	*A*
Cooling by active vaporization	Rapid	*B*
Cooling in liquid	Slow	*C*

For convenience in reference we may call these three modes of cooling *A*, *B*, and *C*, respectively. Cooling in accordance with this series of stages has been found true in all of the liquids examined.[8] With oils, however, the lack of a definite boiling point, resulting in a more or less continuous vaporization to temperatures well below the nominal boiling point, prevents a clear separation of the two latter cooling stages. The use of cylinders of considerably larger size has indicated that both the division of the quenching action into a number of steps and the temperatures delimiting them are quite independent of the size of the object and that the action is a true surface temperature effect.

[8] The use of the quenching liquid as a jet or spray favors the destruction of the thermal structures above described and results obviously in the forced extrapolation of the desirable B-stage both upward, displacing the A-stage, and downward toward the boiling point of the liquid.

Measurements of Quenching Power

The quenching power of a liquid may be defined as its ability to absorb heat rapidly. Evaluation in absolute units would involve the determination of the constants embodied in the exponential laws of cooling obeyed, and the separation of such components of these constants as are associated with the properties of the cooling mass from those that are properties of the liquid alone. For the present purpose, the rate of heat abstraction from the standard cylinder is sufficiently definite and its proportional quantity cooling velocity (degrees per second) will here be taken as a measurement of the quenching power. The quenching power measured in this way is constantly changing during the course of cooling and comparisons of this quantity between two liquids can be made only with the reservation that the reference point is the same in both cases. Thus, the results will be given in a set of curves showing the changes in quenching power of each liquid during quenching at several different liquid temperatures, and comparisons of the quenching power of the different liquids will be made at a temperature of 700° C. of the cooling cylinder.

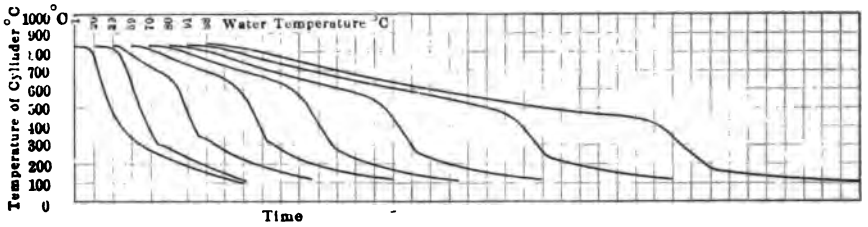


Fig. 8.—Cooling curves of standard cylinder quenched in water; each square represents 1 sec.

Water.—In Fig. 8 is plotted a series of cooling curves of the standard cylinder quenched in water, in which the water temperature was varied from 0° to nearly 100° C. Beginning with the water at 98,° we find the same general features of cooling apparent as in the preceding analysis; *i.e.*, a three-stage operation. As the initial temperature of the water is lowered the steepness of the *A* stage increases and passes over into the *B* stage sooner and at increasingly higher temperatures. For example, in the case of water at 98°, the cylinder cooled in the *A* stage for 23 sec. after immersion and passed into the *B* stage at about 400° C.; with water at 39° the *A* cooling lasted but 3.5 sec. before passing into the *B* stage at 650°. The behavior is intermediate at intermediate temperatures. With the water temperature reduced to 20° C., the distinction between *A* and *B* stages (at 700°) is ill defined; and in water at 0°, the cooling appears to be entirely of the *B* stage from the initial temperature of quenching down.

Quenching power curves are given in Fig. 15. The high quenching power of cold water is due to the fact that cooling begins at once in the *B* stage. As the temperature of the water is raised, the tendency toward the formation of the permanent vapor layer becomes greater and in the

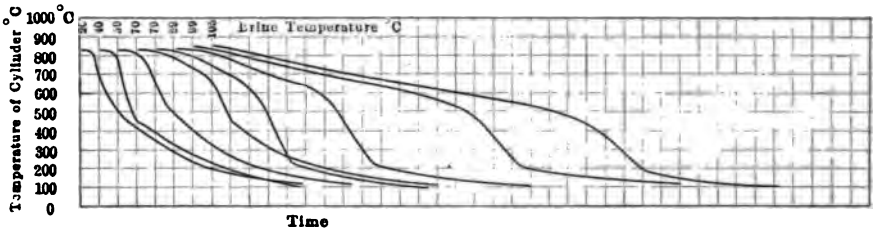


FIG. 9.—COOLING CURVES OF STANDARD CYLINDER QUENCHED IN BRINE; EACH SQUARE REPRESENTS 1 SEC.

curve taken at 39° the initial cooling is distinctly of the *A* type with a correspondingly great impairment of quenching power. With still further increases in temperature, the *A* stage makes further inroads on the

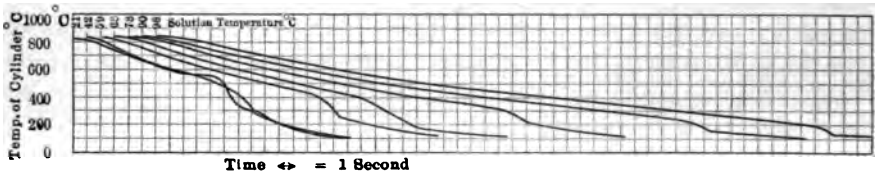


FIG. 10.—COOLING CURVES OF STANDARD CYLINDER QUENCHED IN SOAP SOLUTION.

B stage, the temperature range over which it persists increases and its own heat-absorbing power decreases.

Brine.—Cooling curves are given in Fig. 9 and quenching-power curves in Fig. 16. With cold brine, cooling starts in the *B* stage with a cooling

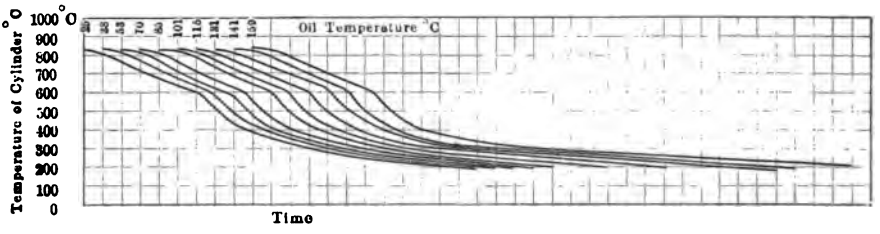


FIG. 11.—COOLING CURVES OF STANDARD CYLINDER QUENCHED IN OIL No. 1; EACH SQUARE REPRESENTS 1 SEC.

power about equal to 0° water. This it maintains up to 60,° when the *A* stage begins to make its appearance. At temperatures above this, brine much resembles water in its quenching behavior.

Soap Solution.—Cooling and quenching-power curves are shown in Figs. 10 and 17. The *A* stage is prominently present at temperatures

as low as 20° when the solution sets into a thick jelly. This gives a very low initial quenching value continuing to cylinder temperatures 200°–300° lower than with water under corresponding conditions.

Oil Number 1.—Cooling curves and quenching-power curves for the first oil bath are given in Figs. 11 and 18. Cooling starts slowly with the *A* stage and passes into the *B* stage, with an attendant sharp increase in quenching power, uniformly at 600° C. for all oil temperatures as high as

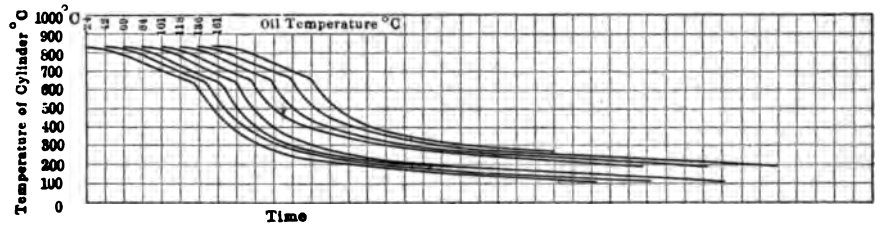


FIG. 12.—COOLING CURVES OF STANDARD CYLINDER QUENCHED IN OIL NO. 2; EACH SQUARE REPRESENTS 1 SEC.

used (160° C.). The *B* stage is but weakly developed and passes almost imperceptibly and with considerable overlapping into the *C* stage at about 400°, resulting in a very low cooling power from this temperature downward. As may be seen from the quenching-power curves, Fig. 18, the heat-absorbing properties of this oil in the *A* stage (800°–600°) and the *B* stage (600°–400°) are practically the same for all oil temperatures be-

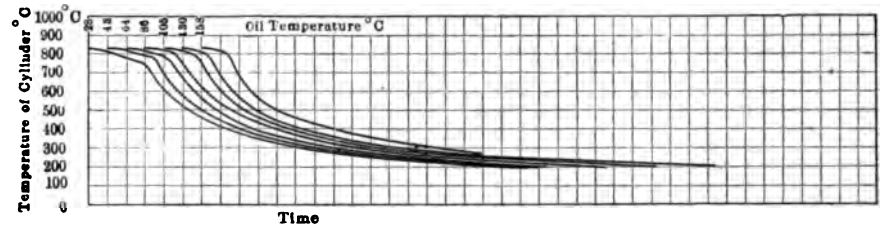


FIG. 13.—COOLING CURVES OF STANDARD CYLINDER QUENCHED IN OIL NO. 3; EACH SQUARE REPRESENTS 1 SEC.

tween 25° and 160° C.; below 400°, in the *C* stage, the cooling power, already quite low, is progressively reduced still more as the temperature rises.

Oil Number 2.—This oil has certain peculiarities in which it differs from No. 1; its curves are shown in Figs. 12 and 19. The initial cooling in the *A* stage terminates at 650° C. in all cases; the quenching power in that range is the same as that of the preceding oil. In the *B* stage we find its cooling power initially about double that of No. 1 oil and maintaining a high value (cooling velocity = 50° per sec.) down to 300°. This condition persists to an oil temperature of about 90,° when an abrupt

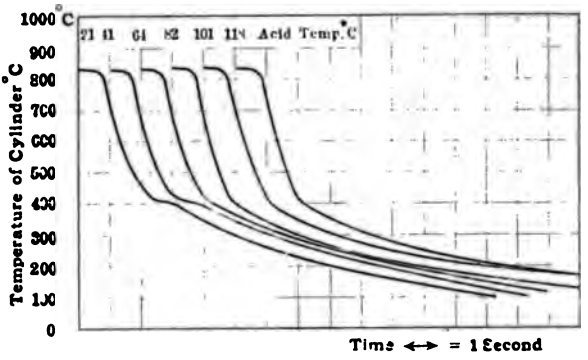


FIG. 14.—COOLING CURVES OF STANDARD CYLINDER QUENCHED IN CONCENTRATED SULFURIC ACID.

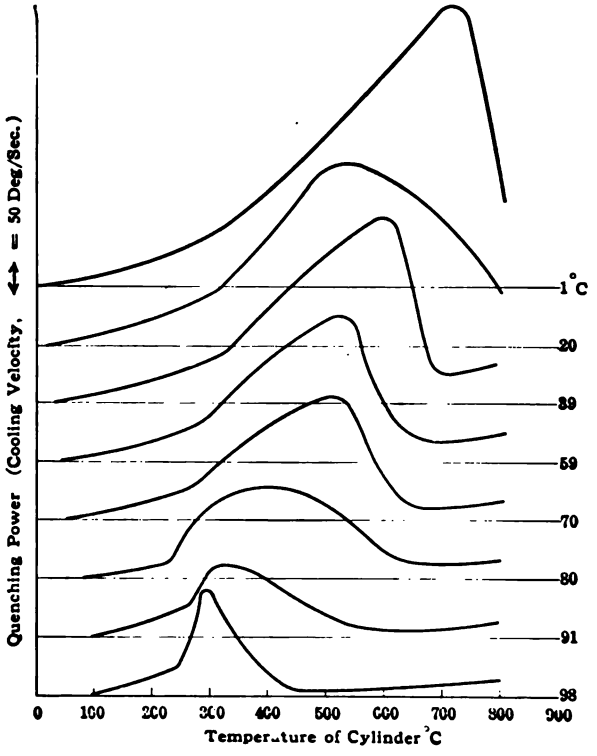


FIG. 15.—QUENCHING POWER OF WATER.

change in the *B* quenching power occurs, decreasing to values nearly identical with those of oil No. 1, with the zone of rapid cooling passing away at about 100° higher. In quenching effect, the combined *B* and *C* stages are shifted abruptly 100° higher at oil temperatures above 90° C.

Oil Number 3.—Figs. 13 and 20. A slight development of the *A* stage may be seen on the cooling curves, terminating between 750° and

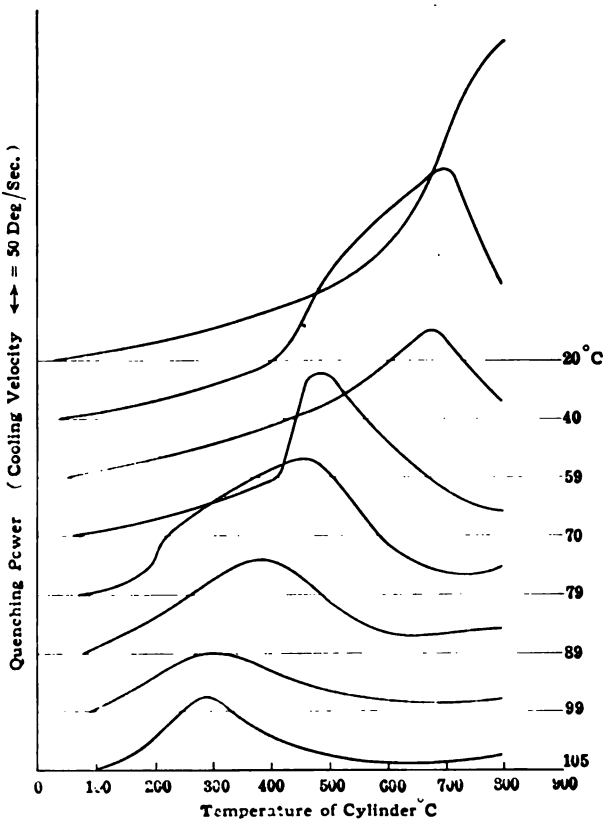


FIG. 16.—QUENCHING POWER OF BRINE.

800° C. Quenching from higher temperatures than this should show a full development of it. The major part of cooling thus lies in the *B* and *C* stages, giving a very considerably greater initial quenching value than is the case with the other two oils. The *B* quenching power is approximately 25 per cent. greater at an oil temperature of 160° than at 28°. It passes into the *C* stage in the temperature range 500°–550°; the cooling powers of the latter are quantitatively similar to oil No. 1.

It may be noticed that the temperature marking the transition from

the slow *A* to the quick *B* stage in these three oils is proportional to the boiling point of the oil.

Concentrated Sulfuric Acid.—As shown in Fig. 14, cooling begins at once in the *B* stage with a very high quenching power, somewhat greater than that of brine, and continues to about 400° C. where *C*-stage cooling, with its characteristically low rate of heat absorption, begins. The quenching-power curves are shown in Fig. 21. The major mechanism

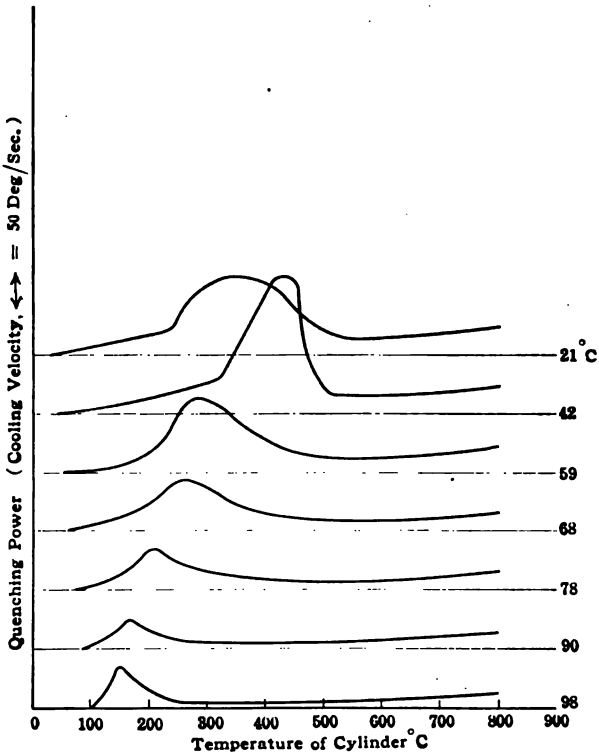


FIG. 17.—QUENCHING POWER OF SOAP SOLUTION.

of quenching is not affected by a rise in acid temperature to 118° C. and at no time is *A*-stage cooling observable. The high *B*-stage quenching power has a tendency to decrease at the higher acid temperatures.

A convenient reference point from which to begin a comparison of the various liquids is the quenching power at 700° C., approximately the temperature of the A_1 transformation in carbon steels on cooling. This information is plotted in Fig. 22. Considering first the aqueous solutions, it will be noticed that the quenching power of brine remains sensibly constant at a high value for a certain increase in the solution temperature, then falls off sharply to a much lower value through an

intermediate range of temperature and is then rather constant at this low value. The first part of the curve corresponds to *B*-stage cooling; the steep portion to the transition from *B*-stage to *A*-stage cooling; and the last portion to fully developed *A*-stage cooling. Water, even at as low a temperature as 1°, is rapidly declining in quenching power due to its lessening powers of resistance against the formation of the fatal vapor jacket. It will be seen that water, brine, and soap solution all

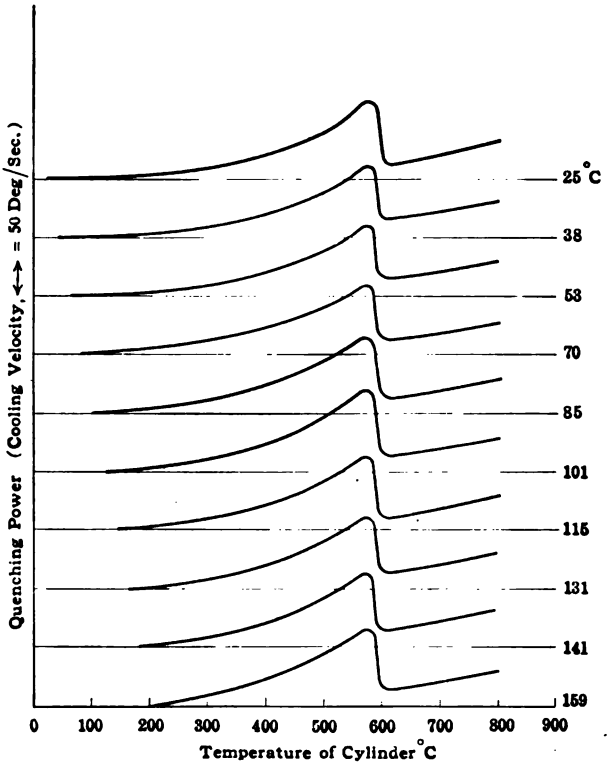


FIG. 18.—QUENCHING POWER OF OIL NO. 1.

blend into a common curve in the *A* zone, indicating that in this stage the cooling agent is the same for all and is, of course, steam.

An interesting check on the conformation of these curves is given by McCance,[9] who found the hardness developed in small pieces of steel of eutectoid composition, by quenching in water and salt solutions as the bath temperature was progressively raised, to be maintained at a constant value up to a temperature of 20° C. for water and 60° C. for brine, beyond which temperatures in both cases the hardness fell to a value one-half as

[9] McCance: A Contribution to the Theory of Hardening. *Jnl.* Iron and Steel Inst. (1914) **89**, 192.

great and decreasing slightly as the liquid temperature was raised to the boiling point.

If a comparative value of 1.0 is taken for the quenching power at 700° in the fully developed *A* stage of an aqueous solution, *i.e.*, 100° solution temperature, then the relative values of the different liquids at 20° is as follows:

COMPARATIVE QUENCHING POWER

Water	11
Brine	20
Soap solution	1.3
Oil No. 1	2.5
Oil No. 2	2.0
Oil No. 3	10
Sulfuric acid	24

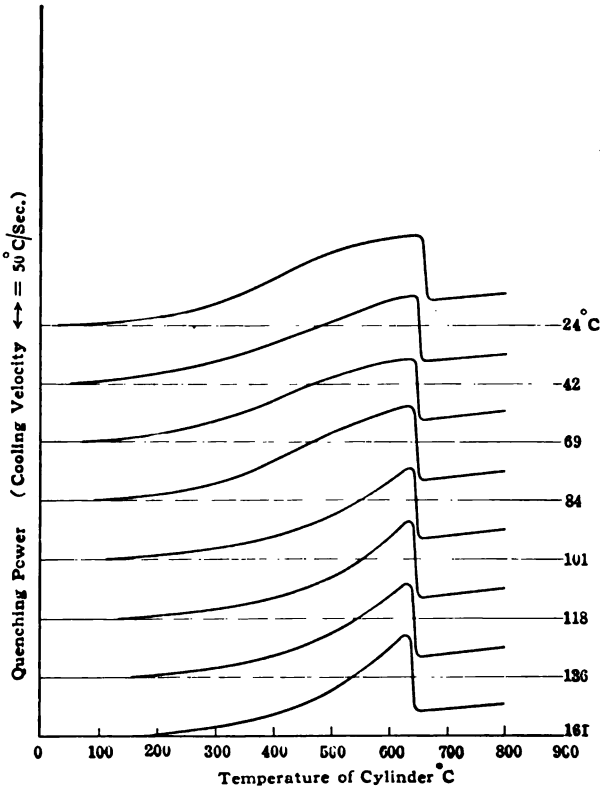


FIG 19.—QUENCHING POWER OF OIL No. 2.

The peculiar behavior of the soap solution in increasing slightly in quenching power up to 30° is in accordance with the nature of the solution; at 20° it is a thick jelly and at 30° it is liquid.

Stability of quenching power with respect to change in liquid tempera-

ture is a valuable and, perhaps, necessary property for a good quenching liquid, and its realization is found when cooling is restricted to any one of the three stages (A, B, or C). A general survey of the quenching properties of the several liquids is given in Fig. 23, in which are plotted, in seven diagrams, the fields in which each of the three cooling stages are active, with respect to the liquid temperature and the temperature of the cooling surface. The line of separation between the A and B fields was taken as that temperature at which an increase in the rate of cooling was first apparent; the actual transition from A to B stages takes

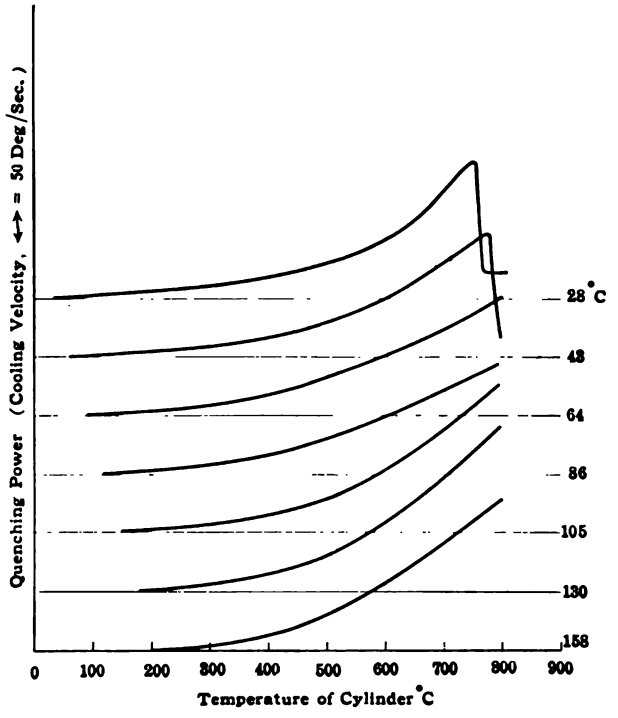


FIG 20.—QUENCHING POWER OF OIL NO. 3.

about 100° to accomplish, and hence too quantitative an interpretation of the diagram is to be avoided. In a diagram of this sort, stability of quenching power is indicated when the horizontal line representing the surface temperature with reference to which the quenching power is taken remains wholly within one field. Conversely, in passing from one field into another changes in quenching power of large magnitude may be expected. Thus the unstable nature of water at 700° is obvious and it can be asserted that its quenching power at a surface temperature of 500° or 400° would be much more stable. Applying the same criterion, brine, while not completely stable at any surface temperature, would have a limited stability with respect to rapid cooling safely to 60°.

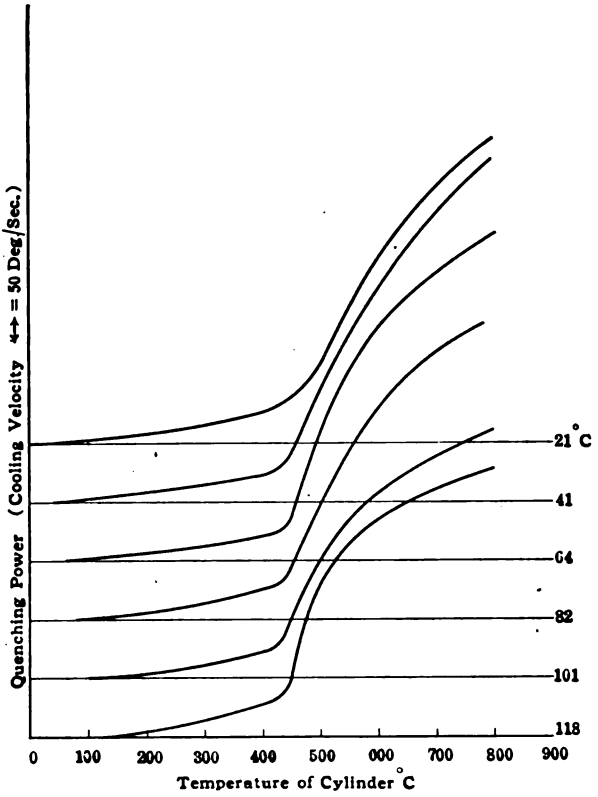


FIG. 21.—QUENCHING POWER OF SULFURIC ACID.

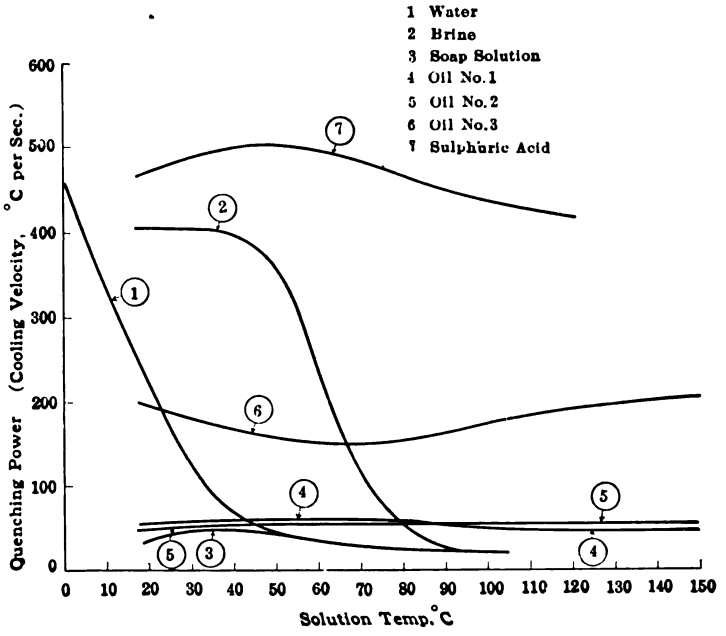


FIG. 22.—QUENCHING POWER OF TECHNICAL QUENCHING LIQUIDS BASED ON COOLING VELOCITY AT 700° C.

At 700° and 600°, the soap solution lies wholly within the *A* field, yielding at these temperatures a slow medium, safe in this respect throughout the whole scale of solution temperatures. Notice the narrow and distorted *B* field, which would be expected to minimize the stresses caused by rapid cooling through a considerable temperature interval. The com-

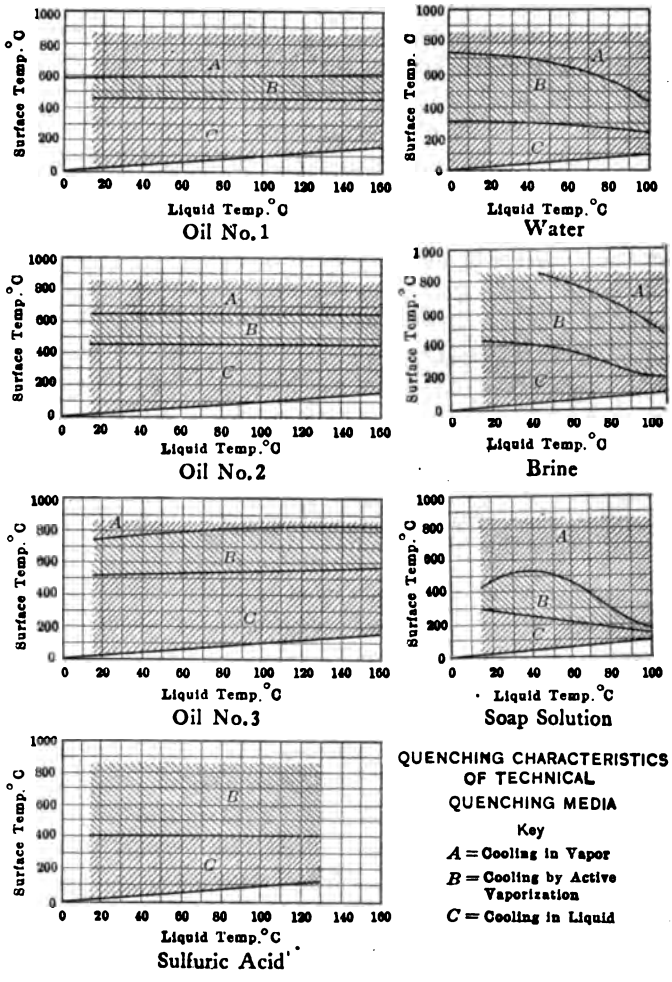


FIG. 23.

plete stability of all the oils and sulfuric acid is indicated by the horizontal orientation of all their fields.

In conclusion it should be emphasized that the measurements given have been made with a small mass, quickly responsive to changes in external conditions, and represent the changes in environment that a large mass may be expected to follow only with considerable thermal inertia.

DISCUSSION

H. M. Howe, Bedford Hills, N. Y.—One wonders whether it might not be well to check, by a rather simple and direct way, the inferences that would be drawn from this investigation. Suppose we quench in these various liquids cylinders of steel, selecting material that is not likely to split in hardening, and then examine their cross-sections microscopically. If the cylinder is of any size, you will find a series of concentric bands, a very hard band, a softer band, and a hard band—but it seems that here would be a way of checking the inferences from the cooling curves by comparing these waves of hardness obtained in similar specimens quenched in different liquids. One would think that there would be great promise in that way.

One's first thought is to quench similar pieces of steel, similarly heated, in different liquids and determine their hardness. I suppose many of us have tried that. We then meet the difficulty that it is not the most rapid cooling that gives the greatest hardness. There is an intermediate rate of cooling that gives maximum hardness so that you can not by any simple method as that check this cooling method. But it seems possible that by studying these waves, by their amplitude, determining the distance you must go to reach a given hardness at the crest of the wave, you might infer the relative rapidity of cooling perhaps even more certainly than you could from the cooling curves themselves.

T. D. Lynch.—After the paper was printed and in circulation we made some tests along the lines suggested by Doctor Howe. We selected a square bar of steel long enough to make 24 samples from the same bar suitable for Izod testing. The chemical analysis of the bar was: carbon, 0.58 per cent.; manganese, 0.38 per cent.; phosphorus, 0.009 per cent.; sulfur, 0.036 per cent.; silicon, 0.12 per cent. Nine of these samples were heated to 830° C. and quenched in water at temperatures, varying by 10°, from 20° to 100° C. The fractures of these test samples are shown at the bottom of Fig. 24, numbering from left to right: No. 1, 20° C.; No. 2, 30° C.; up to No. 9 which was quenched in water at 100° C. The change in the character of the fracture above 60° C. is very marked. Nine samples were heated to 830° C. and quenched in a saturated solution of brine. These are shown in the middle of Fig. 24, beginning at the right with the brine at 20° C. and ending at the left with the brine at 100° C. The change of grain structure takes place at about the same place as that shown by the water. Two samples were quenched in oil at 25° C., two at 70° C., and two were not quenched at all. These fractures are shown at the top of Fig. 24. Beginning at the left, the first two were unquenched, the second two were quenched at 25° C.; and the third two, at 70° C. The fractures caused by the quenching in oil at 25° C. and 70° C. are very similar.

These tests show very clearly that the grain structure of the specimens was affected much more with cold water or brine than with hot water or hot brine and that the temperature of the quenching oil does not seriously affect the grain structure. They also show that the effect of the cold water and cold brine is more drastic than oil at any temperature; that hot water or hot brine may be so controlled as to give practically the same, or less, effect than that of oil as a quenching medium; and that hot water and hot brine give a less drastic treatment than oil at any temperature.

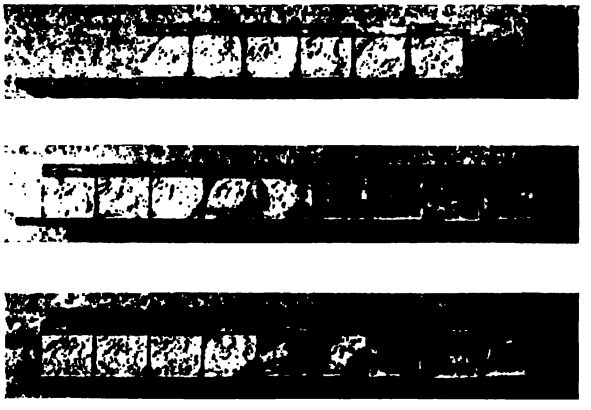


FIG. 24.

RALPH M. SHERRY,* Detroit, Mich.—One item has been omitted in some of these papers; that is the speed of the quenching medium or the volume applied. With a large volume of oil it is possible to obtain the same results as with water. It seems to me that in conducting work of this kind some attention should be paid to that factor. From years of practical experience, I know that perfect results can be obtained with oil if it is given a high rate of speed and I wonder if there has been any work done on this that may be brought out in discussion.

T. D. LYNCH.—These tests were made without agitating the oil or the water, which we believe is the most common practice, especially in the case of large forgings, and, while it is possible to quench steel parts in a moving liquid or by moving them through the liquid, these tests were carried out with the idea of having them represent common practice rather than possibilities. I appreciate fully the fact that flowing oil makes a decided increase in the quenching value of oil; rapidly moving air also has a marked difference on the cooling of a piece of steel in the air. We have found from experience that if you cool a piece of steel quickly by passing a strong current of air over it, you will get a true elastic

* Chief Metallurgist, General Motors Corpn.

limit perhaps double that of a similar piece of steel cooled slowly in the furnace and I have found in some of our work that it has been possible to double the elastic limit, raise the tensile strength considerably, and, at the same time, increase the ductility. These are possibilities which may bear some fruit in commercial work as time goes on.

R. M. SHERRY.—In our practice it is essential to assure a large flow of moving liquids. It is impossible to harden the work according to specifications without an extreme flow of water or by quenching vertically in the tank. In ordinary commercial practice, it is absolutely essential to distinguish between moving liquids and still liquids.

THE CHAIRMAN (J. A. MATHEWS, Syracuse, N. Y.).—One point that has not been touched upon is the permanence of quenching mediums. How long can they be used without suffering deterioration? That must be looked after very carefully in quenching oils, uniform results cannot be obtained with oils that deteriorate with repeated use. Another point is the necessity of going very slowly in adopting conclusions from papers of this kind without considering the different steels and their capacity for being hardened. The speed of quenching is one thing and the capacity for hardening is quite a different thing, so while the oil or liquid might be perfectly satisfactory for one kind of steel, it may not be adapted to some other steel of different composition and different capacity for hardening.

NORMAN B. PILLING.—We are quite aware of the limitations of the paper, which are discussed in its introductory paragraphs. With respect to moving liquids, it is quite obvious that the effect is simply the tearing away of the vapor jacket surrounding the cooling body, resulting in the forced extrapolation of the zone of rapid cooling (the B stage) to temperatures higher than that at which it normally exists.

In regard to Doctor Howe's suggestion that another line of attack would be furnished by a microscopic examination of steel quenched under various conditions, it is apparent that pieces of considerable size would be required, necessitating a quantitative knowledge of the internal temperature gradients during quenching. We find that in a steel cylinder of the dimensions of a lead pencil the microstructure transversely is quite uniform, except at the critical cooling speed of the steel when small differences in cooling velocity exert a relatively large effect on the completeness of martensitization. In this case the proportion of martensite to troostite does change in a radial direction.

Effect of Rate of Temperature Change on Transformations in an Alloy Steel*

BY H. SCOTT,† A. B., WASHINGTON, D. C.

(New York Meeting, February, 1919)

SINCE Böhler discovered, in 1903, on cooling certain alloy steels, the phenomenon of a new and lower temperature transformation than the usual Ar_{3-2-1} obtained by increasing the maximum temperature to which the material was heated, much has been published[1] connecting this phenomenon with a large number of dissimilar steels of high alloy content. From the fact that the transformation divides itself, taking place at two widely separated temperatures, it has been called a "split transformation." The significant facts established by recent investigators[2] are: that when the transformation occurs at the higher temperature Ar', troostite or a decomposition product is formed, and that when the transformation occurs at the lower temperature Ar'', the resulting structure is martensite. The terminology Ar' and Ar'' adopted here is that of Portevin.[3]

PREVIOUS INVESTIGATIONS

Reviewing the work on this subject published in a recent issue of *Revue de Métallurgie* and referring in particular to his statement that martensite is a solution of carbide in alpha iron, Le Chatelier says:[4]

"How then can a theory already 20 years old demand new investigations? The reason for it is that we have not succeeded in proving directly the real presence of the transformation of iron during the very short duration of the quenching. The fall of temperature takes place at the rate of several hundred degrees per second and the observation of

* Contributed through the courtesy of the U. S. Bureau of Standards.

† Assistant Physicist, U. S. Bureau of Standards.

[1] Yatsevitch: Recherches sur l'Acier à Coupe Rapide. *Rev. de Met.* (1918) **15**, 65–115. Bibliography to 1915.

[2] P. Dejean: Les Points Critiques de Refroidissement des Aciers Auto-Trempants et la Formation de la Troostite et de la Martensite. *Rev. de Met.* (1917) **14**, 641–675.

A. Portevin: Les Points de Transformation des Aciers Nickel-Chrome. *Rev. de Met.* (1917) **14**, 707–716.

C. A. Edwards: Initial Temperature and Critical Cooling Velocities of a Chromium Steel. *Jnl.* Iron and Steel Inst. (1916) **93**, 114–140.

[3] Portevin, *Loc. cit.*

[4] Henri Le Chatelier: La Trempe de l'Acier. *Rev. de Met.* (1917) **14**, 601–606.

phenomena so rapid requires particularly sensitive methods of recording. I have attempted without success to observe the moment of the reappearance of the magnetic property during the quenching of bars 15 mm. square, but the inequalities of temperature from one point to another in the mass conceal the phenomenon. M. Chevenard,[5] in working on wires of a diameter 100 times smaller and using as a characteristic of the transformation of iron the change of length instead of the variation of magnetism, has surmounted for the first time the difficulties which seemed at first view insurmountable and he has done it with an extreme precision. The thermal measurements of Portevin and Garvin[6] and of Dejean[7] lead to the same conclusions, although in a less direct manner."

The results presented here add further confirmation of the theory of Le Chatelier though in a less direct manner than those of Chevenard. Thermal and microscopic data are brought forward here to establish the effect of rate of temperature change on the temperature and nature of the transformations in a steel of the composition: carbon 1.75, manganese 0.26, cobalt 2.90, chromium 15.0. Similar work has been done by Edwards[8] on a steel of the composition: carbon 0.63, chromium 6.15, silicon 0.07, manganese 0.17. He arrives at the conclusion, however, that "the maximum hardness was obtained when the thermal transformation had been entirely prevented, and when this was accomplished the steel was purely martensitic in structure."

The present work fails to confirm this statement, as does that of the investigators already referred to. Edwards was unable to observe the transformation Ar'' with the formation of martensite, probably for the reasons given by Rosenhain[9] in his discussion of Edwards' paper.

Yatsevitch,[10] Dejean,[11] and Honda[12] have used two or three cooling rates in their experiments with varying maximum temperatures and their results show, as do Edwards', that the transformation split occurs for lower values of the maximum temperature with faster cooling rates.

The previous investigators in this field have laid particular stress on the variation of the maximum temperature, the rate remaining

[5] Pierre Chevenard: Dilatomètre Différential Enrégistreur. *Rev. de Met.* (1917) **14**, 610–640.

[6] Portevin and Garvin: Influence de la Vitesse de Refroidissement sur la Température de Transformation de la Structure des Aciers au Carbone. *Rev. de Met.* (1917) **14**, 607–609.

[7] P. Dejean: Les Points Critiques de Refroidissement des Aciers Auto-Trempants et la Formation de la Troostite et de la Martensite. *Rev. de Met.* (1917) **14**, 641–675.

[8] C. A. Edwards, *Loc. cit.*

[9] Walter Rosenhain: Discussion on Initial Temperature and Critical Cooling Velocities of a Chromium Steel. *Jnl.* Iron and Steel Inst. (1916) **93**, 147.

[10] Yatsevitch, *Loc. cit.*

[11] P. Dejean, *Loc. cit.*

[12] Kôtarô Honda and Takejirô Murakami: On the Structure of Tungsten Steels and Its Change under Heat Treatment. *Sci. Rep.* Tohoku Imp. Univ. (1917) **6**, 235–283.

constant, while the variation of rate, the maximum temperature remaining constant, has received little attention. The present work attempts to apply the latter method to the investigation of an alloy steel with the object of correlating the results of that method with those of the former and to establish the relations of the several phenomena observed.

Experimental Method

The method employed for obtaining the thermal curves was to heat the samples attached to the hot junction of a 0.5-mm. diameter platinum wire, 90 platinum − 10 rhodium thermocouple, in an electric vacuum furnace, taking potential measurements on a dial potentiometer and measuring the time interval on a chronograph, as described in Bureau of Standards *Scientific Paper* 213. The furnace, which was recently built at the Bureau, is a modified type of the one described by Rosenhain[13] as in use at the National Physical Laboratory. It is admirable for the purpose, as extreme rates of temperature change can be obtained with smooth curves over long ranges.

Thermal Curves

The curves of Figs. 1 and 2 were plotted by the inverse-rate method from readings taken every 0.02 millivolt (approximately 2° C.) except for several extremely fast runs, which, however, are plotted on that basis. The curves of Fig. 1 are a preliminary series taken on sample A of about 10 gm. mass to locate the transformation ranges and without fully knowing the characteristics of the material. The data for the curves of Fig. 2 were taken on sample B, mass 0.81 gm., keeping the maximum temperature constant and extending the observations to lower temperatures than for sample A. The values given for the rate of temperature change were reduced from the inverse-rate curve observations taken on heating just before $Ac_{1\text{-}3}$ and on cooling midway between Ar' and Ar''.

The transformations, as designated on the curves of Figs. 1 and 2, are Ac_t, an evolution of heat on heating a sample previously cooled at a rate that gave Ar''; Ac_2, the magnetic transformation; $Ac_{1\text{-}3}$, the transformations Ac_1 and Ac_3 merged or nearly superimposed; Ar', the upper transformation of the split Ar transformations; and Ar'', the lower transformation. The temperature values of these transformations are collected in Table 1.

The appropriateness of the transformation notation $Ac_{1\text{-}3}$ and Ac_t will be seen from the discussion of those transformations. Two values for the maximum transformation temperature indicate a double peak.

[13] Walter Rosenhain: Some Appliances for Metallographic Research. *Jnl.* Inst. Met. (1915) **13**, 164–172.

(See H. Scott and J. R. Freeman, Jr.: Use of Modified Rosenhain Furnace for Thermal Analysis. *Pyrometry Volume* (1920) 214.

In Fig. 3, the temperature values of Ac_{1-3}, Ar', and Ar'' given in Table 1 are plotted against rate of temperature change in degrees C. per second.

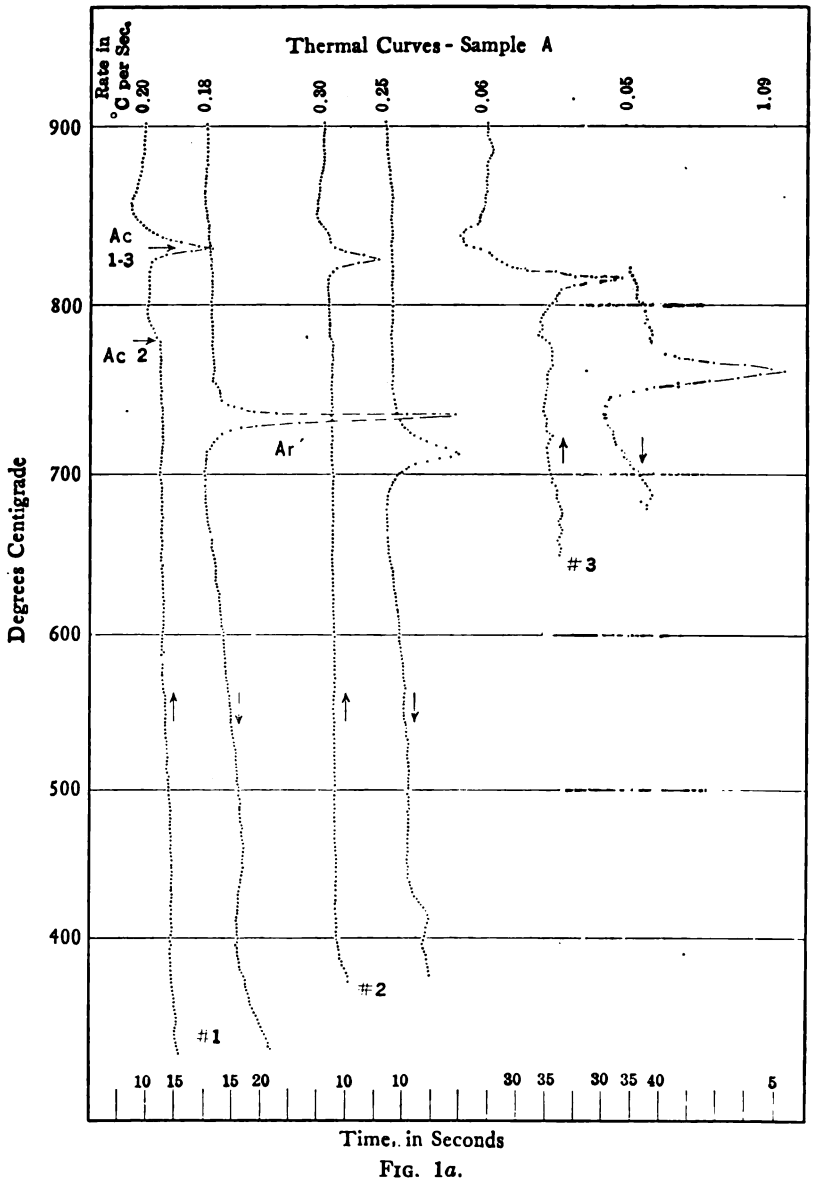


FIG. 1a.

No attempt is made to interpret the double peaks and lines representing Ar' and Ar'' in Fig. 3 are rather arbitrarily drawn through the higher values.

Effect of Cooling Rate

An inspection of the cooling curves of Figs. 1 and 2 shows that Ar' is the normal $Ar_{3\text{-}2\text{-}1}$ of slow cooling rates, but that it gradually dies off

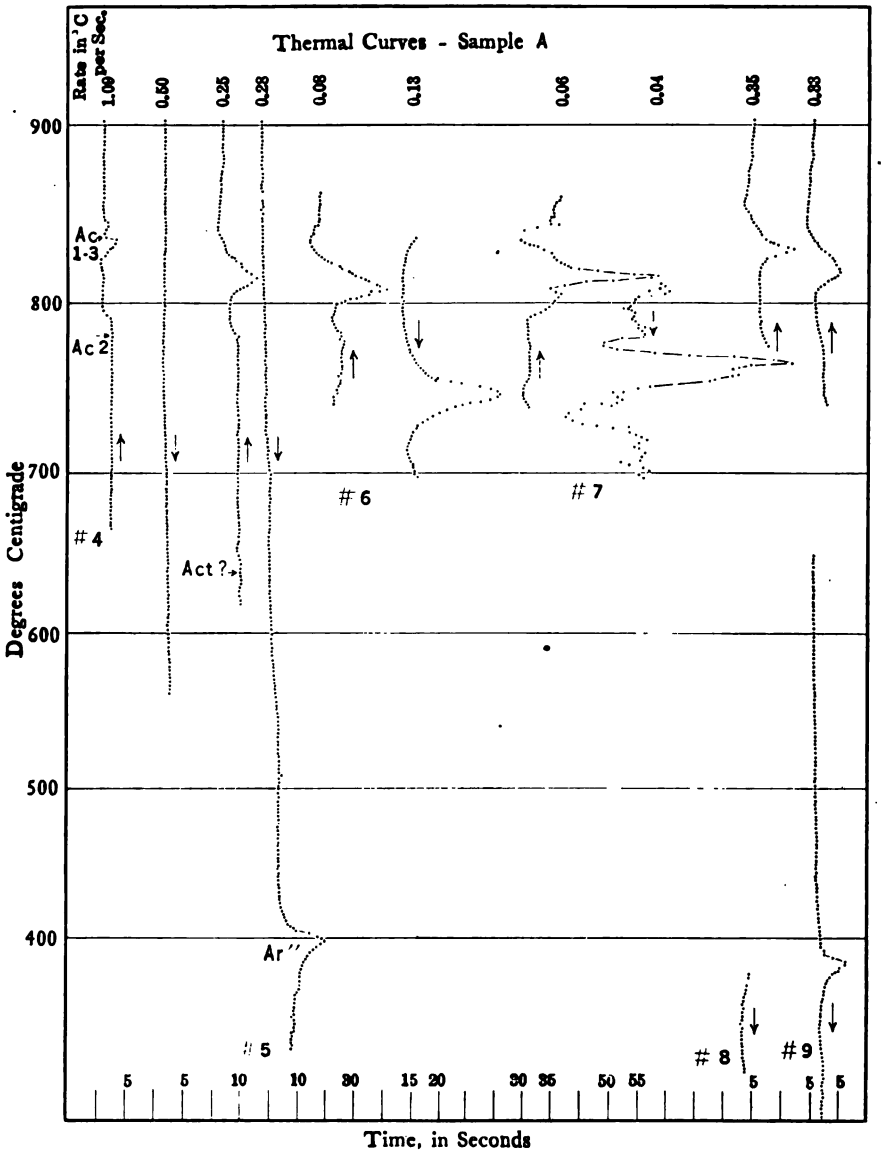


Fig. 1b.

in intensity with increasing rate. While Ar' is falling off in intensity, the transformation Ar'' comes into existence and gains in intensity,

being a maximum for rates that do not show Ar'. This region over which both Ar' and Ar'' occur, as shown in Fig. 3, will be called the critical

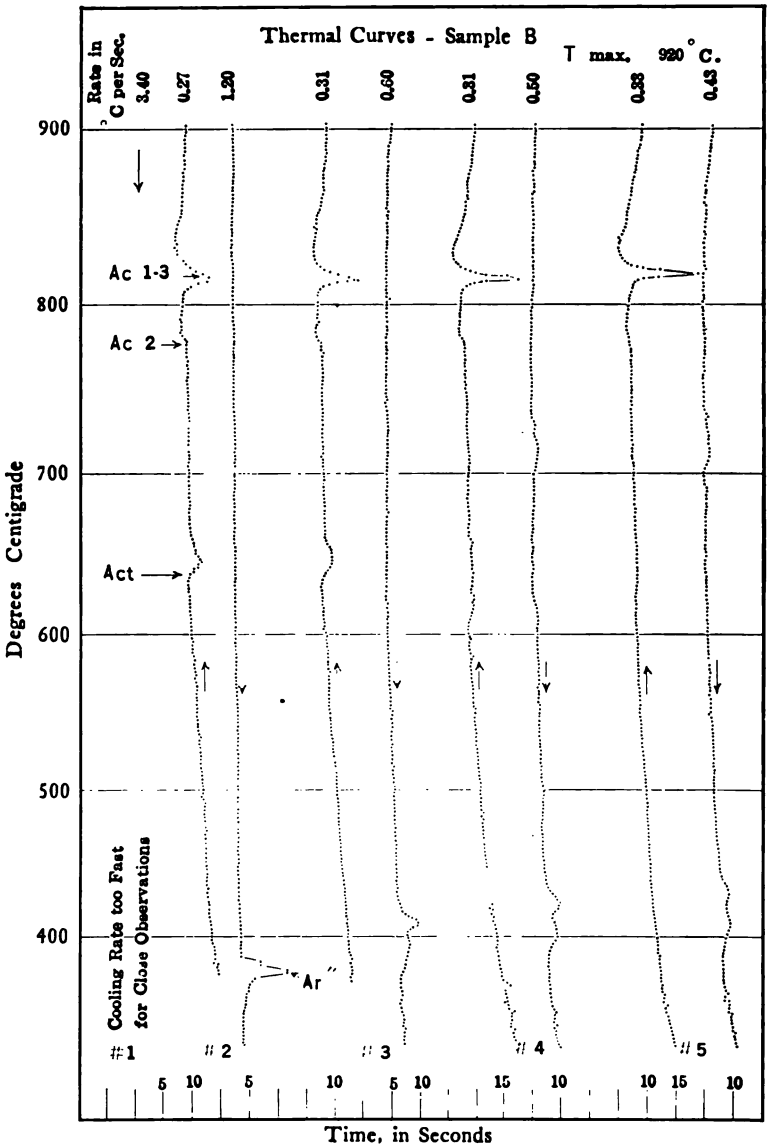


FIG. 2a.

cooling range. Its limits were roughly determined as 0.15° and 0.70° C. per second, by plotting a measure of the transformation intensities obtained by a method to be described in a subsequent section, against rate and extending a straight line through the values back to zero.

The remarkable change in properties caused by this very slight change in rate is represented, when the same phenomenon is observed

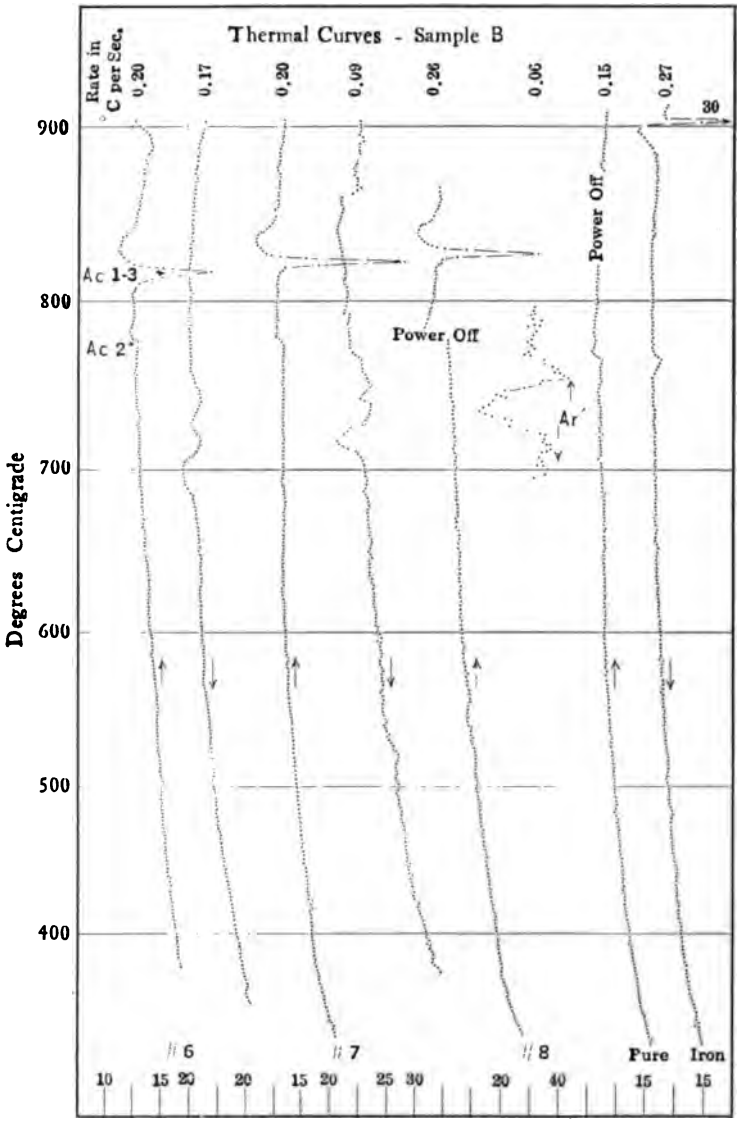


FIG. 2b.

on varying the maximum temperature, by the considerable temperature variation of approximately 300° C. for some high-speed steels.[14] The

[14] Honda and Murakami, *Loc. cit.*
Carpenter, *Loc. cit.*

TABLE 1.—*Temperature Ranges of Transformations, in Degrees Centigrade*

Run	Rate of Heating	Ac_1			Ac_3	Ac_{1-3}			Rate of Cooling	Ar'			Ar''			Maximum Temperature
		Beg.	Max.	End		Beg.	Max.	End		Beg.	Max.	End	Beg.	Max.	End	
Sample A																
1st	0.20	...	...	...	781	825	834	856	0.18	757	735	...	...	...	...	1050
2d	0.30	...	...	...	781	819	829	855	0.25	742	715	691	446	414	395	1125
3d	0.06	...	...	...	780	...	816	841	0.05	...	760	740	...	...	...	1110
4th	1.09	...	...	...	782	...	836	...	0.50	...	...	...	...	...	...	970
5th	0.25	...	...	...	781	804	816	847	0.28	755	...	670	422	396	357	925
6th	0.08	...	...	...	780	794	809	836	0.13	772	748	714	...	...	...	880
7th	0.06	...	...	...	...	809	817	841	0.04	776	766	734	...	...	...	870
8th	0.35	...	...	...	778	821	831	854	0.80	...	...	...	...	...	...	925
9th	0.33	...	...	...	...	809	819	847	0.71	...	...	...	401	385	318	925
Sample B																
1st		Rate of cooling too fast for close observations							3.40	...	...	...	...	...	...	920
2d	0.27	592	637	644	780	808	818	840	1.20	...	...	...	381	373	340	920
3d	0.31	578	634	644	780	809	816	832	0.60	748?	706	681	429	(409)	371	920
4th	0.31	544	601	628	779	808	816	832	0.50	740	711	681	442	(394) (423) (405)	387	920
5th	0.33	518	621?	637	778	813	818	835	0.43	748	(719) (694)	680	449	(430) (410)	392	920
6th	0.20	564	601?	628	778	807	818	829	0.17	759	(743) (719)	700	...	...	...	920
7th	0.20	...	...	...	778	814	824	835	0.09	775	(751) (736)	723	...	...	...	920
8th	0.26	...	...	...	...	821	829	842	0.06	770	756	738	...	...	...	920

fact that the split transformation occurs with a constant value of the maximum temperature shows that it is unnecessary to hypothecate a dissociation of the carbide (or carbides) to explain this phenomenon.

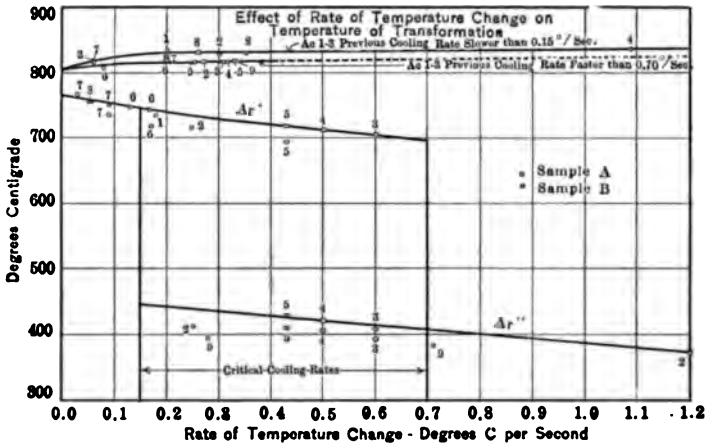
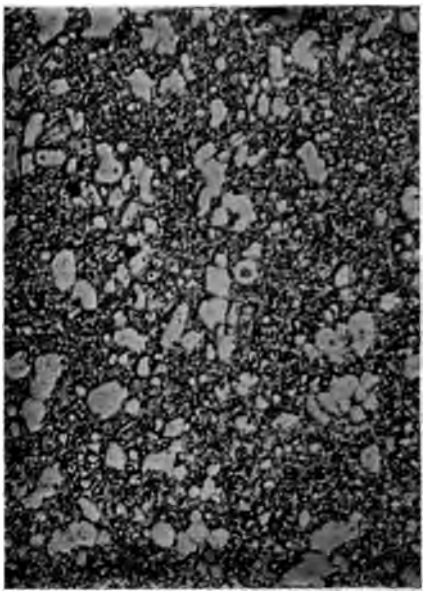


FIG. 3.



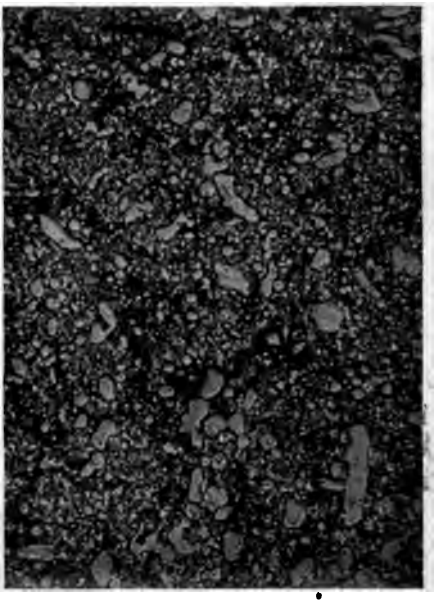


FIG. 4. FIG. 5.

FIG. 4.—COOLING RATE 0.01° C. PER SECOND, TRANSFORMATION Ar'.
FIG. 5.—COOLING RATE 0.30° C. PER SECOND, TRANSFORMATION Ar' AND Ar''.

To establish the structural difference between the material cooled at a rate that gave Ar' and one that gave Ar'' and the analogy to the phenomena obtained by varying the maximum temperature for this steel, micrographs were taken of samples cooled at several definite rates of

cooling.[15] The micrograph, Fig. 4, taken after cooling at a rate of 0.01° C. per second, A_1' only occurring, shows an irregular mass of fine carbide particles corresponding to pearlite in carbon steels and distinct from the coarse particles of free carbide in a ferrite matrix. Figs. 5 and 6, micrographs taken of samples cooled at rates of 0.30° and 0.33° C. per second respectively, show characteristic black troostite patches on a background of martensite. With those cooling rates, the transformations Ar' and Ar'' were both obtained. Fig. 7, which is a micrograph of sample A following a cooling rate of 0.71° C. per second, shows a martensitic

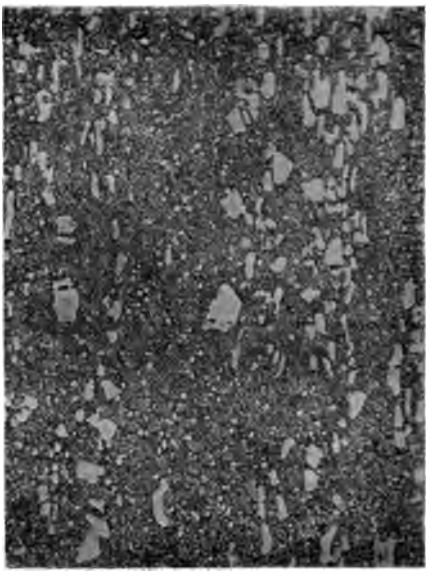


FIG. 6. FIG. 7.

ETCHED IN 2 PER CENT. HNO_3 IN ALCOHOL. × 750.

FIG. 6.—COOLING RATE 0.33° C. PER SECOND, TRANSFORMATION Ar' AND Ar''.
FIG. 7.—COOLING RATE 0.71° C. PER SECOND, TRANSFORMATION Ar''.

structure although the needle-like markings, characteristic of high-carbon steels, are only slightly evident. The conclusions to be drawn from this microscopic evidence are that troostite or a decomposition product forms with the transformation Ar' and martensite with the transformation Ar'', precisely what obtains when the same transformations are observed in other alloy steels with varying maximum temperature.

The radical structural difference between the material showing Ar' and that showing Ar'' presumes a similar radical difference in the transformations Ar' and Ar''. To demonstrate the possibility of this difference, the intensities of the transformations $Ac_{1\,3}$, Ac_2, Ar', and Ar'' have been estimated by means of a planimeter measuring the area of the positive departure of the thermal curves from the assumed neutral body

[15] Micrographs by H. S. Rawdon.

curves through the respective transformation ranges. The results, given in Table 2, show a well marked loss in intensity of the sum of the areas of Ar' and Ar'' at the cooling rate of 1.20° C. per second, which gives Ar'' above. On the assumption that Ar'' is no new transformation other than $Ar_{3,\ 2\ or\ 1}$, the conclusion is that some one or more of the transformations $Ar_{3,\ 2\ and\ 1}$ constituting Ar' is suppressed.

TABLE 2.—*Areas of Thermal Curves, in Square Millimeters, Corresponding to Heat Effects of Transformations in Sample B*

Run	Ac_{1-3}	Ac_3	$Ac_{1-3}+Ac_3$	Cooling Rate, in Degrees per Second	Ar'	Ar''	Ar'+Ar''
1st	..	..	...	3.40			
2d	36	74	110	1.20	...	84	84
3d	40	66	106	0.60	26	72	98
4th	40	68	108	0.50	38	60	98
5th	64	64	128	0.43	56	48	104
6th	60	64	124	0.17	100	..	100
7th	72	72	144	0.09	120	..	120
8th	72	..	...	0.06	116	..	116
Average	..	68	120	....			

The preceding conclusion agrees with the generally accepted conception that martensite is a solid solution of cementite in some form of iron. This means that Ar_1 is suppressed with the formation of martensite and further evidence is not wanting. The transformation intensities indicate that a heat effect of the magnitude of Ar_1 is missing at Ar''. The work of Honda on the magnetic properties of tungsten steels, in a paper before the Iron and Steel Institute, however, shows that the carbide is retained in solution at Ar'' for the carbide in solution does not undergo the transformation A_0. The transformations Ac_1 and Ac_{1-3} offer still further substantiation to which attention will be called in their discussion.

There still remains the possibility that one of the other transformations $Ar_{3\ or\ 2}$ is suppressed. This, however, is manifestly impossible, for A_3 and A_2 coincide when A_3 is depressed[16] below the normal temperature of A_2 and martensite is magnetic. The magnetic curves of Honda and Murakami[17] taken on a number of tungsten steels showing a split transformation with increasing maximum temperature, also indicate the occurrence of A_2 at Ar''.

The conclusion that must therefore be adopted is that Ar_1 is suppressed with the formation of martensite, or that Ar'' constitutes the transformations Ar_3 and Ar_2.

[16] Kôtarô Honda and Hiromu Takagi: On the Cause of the Irreversibility of Nickel Steels. *Sci. Rep.* Tohoku Imp. Univ. (1917) 6, 324.

[17] Honda and Murakami, *Loc. cit.*

Transformations on Heating

The thermal curves of Figs. 1 and 2 show two transformations $Ac_{1\text{-}3}$ and Ac_2 occurring uniformly within narrow temperature limits and a transformation Ac_t occurring only following certain cooling rates. The identity of Ac_2 is established by its markedly characteristic shape and its uniform occurrence at about 780° C., which is in close proximity to its maximum, 768°, in pure iron. This phenomenon of Ac_1 occurring above Ac_2 in alloy steels is not new and has been well established by Moore[18] for a chromium steel. The transformation $Ac_{1\text{-}3}$ hardly needs identification, though attention should be called to its sluggish ending, which indicates that the transformations concerned do not completely coincide. This is further illustrated by the change in area of the peak, which is evidently Ac_1 from the effect of previous cooling rate on its position, with the temperature of its occurrence.

The transformation Ac_t is indicated by an inflection to the left, which denotes an evolution of heat on the heating curve and occurs over a considerable temperature range. It is a maximum following cooling rates that give Ar'' alone, and loses in intensity following decreasing rates through the critical cooling range, becoming zero when Ar' alone occurs. It is therefore roughly proportional in intensity to Ar'' or the amount of martensite present. By its analogy to tempering, the conclusion may be drawn that Ac_t represents the precipitation of the carbide in solution to form at first troostite and as it progresses the coarsening of the carbide.

This phenomenon of a heat evolution on heating steels that show Ar'' was observed by Carpenter[19] on differential thermal curves with which the maximum temperature was varied and connected with tempering.

The nature of Ac_t, a gradual building up of the heat evolution over a long temperature range, may throw some light on the spontaneous heat evolution and also the change in other physical properties of quenched steels, as observed by Hadfield and Brush,[20] by Matsushita,[21] and by Campbell.[22] The indications are that the transformation starts to a minute degree at very low temperatures, possibly at ordinary temperatures, particularly in carbon steels which temper at lower temperatures than alloy steels. The existence of Ac_t as an evolution of heat following

[18] Harold Moore: The A_2 Point in Chromium Steel. *Jnl.* Iron and Steel Inst. (1910) **81**, 268–286.

[19] H. C. H. Carpenter: Types of Structures and the Critical Ranges on Heating and Cooling of High-Speed Tool Steels under Varying Thermal Treatment. *Jnl.* Iron and Steel Inst. (1905) **67**, 433–473.

[20] Charles F. Brush and Sir Robert A. Hadfield: Spontaneous Generation of Heat in Recently Hardened Steel. *Proc.* Royal Soc. of London (1917) **93**, 188–211.

[21] T. Matsushita: On the Slow Contraction of Hardened Carbon Steels. *Sci. Rep.* Tohoku Imp. Univ. (1918) **7**, 43–52.

[22] E. D. Campbell: Rate of Change at 100° C. and at Ordinary Temperatures in the Electrical Resistance of Hardened Steel. *Jnl.* Iron and Steel Inst. (1918) **98**, 421.

cooling rates that give Ar'' is further confirmation of the suppression of Ar_1 with the formation of martensite.

It will be seen on examining Fig. 3 that practically all the temperature values for the maximum of Ac_{1-3} lie on two smooth curves. The data of Table 1 show that the runs which correspond to the numbers on the upper curve were obtained following cooling rates that gave Ar' predominant and those on the lower curve following cooling rates that gave Ar'' predominant. The temperature interval, 10 to 15° C., between those two curves may therefore be attributed to the state of division of the carbide resulting from the previous heat treatment.

The phenomenon noted in the preceding paragraph offers still further substantiation of the suppression of Ar_1 with the formation of martensite. It may be of interest to note that the curves of Fig. 3 drawn through the temperature values of Ac_{1-3} and Ar' do not point toward a common equilibrium temperature Ae_1.

Summary

The results of previous investigation have been taken to show that with the occurrence of a split transformation on cooling alloy steels from increasingly higher temperatures: (1) when the higher temperature transformation Ar' is observed with low values of the maximum temperature, troostite or a decomposition product results, and (2) when the lower temperature transformation Ar'' is observed with high values of maximum temperature, martensite is the resulting product. The present investigation has shown for a certain alloy steel that on varying the rate of cooling, the maximum temperature remaining constant, a strictly analogous phenomenon is observed, increasing rate of cooling having the same effect as increasing the maximum temperature. Conclusions are drawn to the effect that:

1. The transformation Ar' consists of the transformations Ar_3, Ar_2, and Ar_1.

2. The transformation Ar'' consists of the transformations Ar_3 and Ar_2.

3. The transformation Ar_1, suppressed when Ar'' is observed, occurs on heating as Ac_t with an evolution of heat and the formation of troostite or a coarser condition of the carbide.

4. The maximum of the transformation Ac_{1-3} occurs at a higher temperature when the previous cooling rate gave Ar' than when it gave Ar''.

The author desires to express his indebtedness to Mr. H. S. Rawdon for the micrographic work and to Miss P. L. Thompson for her skilful assistance in preparing the experimental data.

DISCUSSION

Samuel L. Hoyt,* Minneapolis, Minn.—I was privileged to see the apparatus used by Mr. Scott, which is quite unique. It is, as the author says, a modification of the apparatus designed by Doctor Rosenhain, but

* *Associate* Professor of Metallography, University of Minnesota.

it is evidently an improvement. The author is to be congratulated on the fact that he has given us, with his apparatus, a quantitative thermal analysis of these critical points. I notice the author refers to Ac_3, Ac_2, and Ac_1. His designation of the Ac_2 point is all right, but naturally in a steel with as much carbon as this, which would mean a hypereutectic steel, the other point might better be considered as an Ac_1 point with possibly another point above the Ac_1 point. Another important point brought out is the fact that, in Fig. 3, the lines representing the Ar_1 transformation and the Ac_{1-3} transformation do not intersect at the zero rate curve, which brings up the question of the hysteresis requirements of allotropic transformation. It has been stated recently that if we have a true allotropic transformation we must have hysteresis. Referring to the earlier work of the Bureau of Standards on this point, in the paper by Messrs. Burgess and Crowe,[23] we know that as the rate of cooling decreases, and as the mass decreases toward zero, the Ac_3 and Ar_3 points approach each other but do not become equal. At the equilibrium temperature between the high and low temperature modifications, the tendency of one allotrope to change into the other allotrope is practically zero. This, of course, means that even though we reduce both the rate and the mass to zero, we do not necessarily cause the Ac_3 and the Ar_3 points to coincide. In the present case, we have on cooling a change from a solid solution to a complex, and on reheating we have the recombination of the complex into the solution.. The two cases are not identical, but the laws governing them are very similar in many respects. At any rate, we should not expect, even for zero rate of cooling, to find the points on heating and on cooling to coincide. I believe that that is quite in line with the hysteresis requirement of allotropy.

The author refers to the precipitation of carbide from the solution as the cause of the Ac_t point. I quite agree with him as to the mechanism that produces the Ac_t point, but I wish to caution against accepting too readily, or at least using, the mechanism of that transformation as an indication of the constitution of the solid solution of the martensite or the austenite, whichever it is. If ice separates out of a brine solution, we don't say that the water in the brine was in the form of ice. In like manner, we should not say that because the carbide comes of the solution the carbon in solution was that of the carbide. At present our information on the nature of these solutions (austenite and martensite) is not sufficient to permit definite conclusions to be drawn.

H. SCOTT.—I may say that the transformation I called Ac_{1-3} corresponds to the eutectoid transformation obtained on heating a high-carbon steel minus the Ac_2 transformation. I have used this nomenclature for the sake of clarity to show that the two transformations occur together, although it may not be good form to do so.

[23] *Trans.* (1913) **47**, 665.

Metallographic Investigation of Transverse-fissure Rails with Special Reference to High-phosphorus Streaks

BY G. F. COMSTOCK,* A. B., MET. E., NIAGARA FALLS, N. Y.

(New York Meeting, February, 1919)

THE subject of transverse fissures in steel rails has been discussed very thoroughly in recent years from various points of view and the final opinions expressed may be roughly classified into two groups:

That these fissures are the result merely of fatigue of the steel and are independent of the quality of the metal.

That the quality of the metal and the mill practice must have something to do with them.

The first point of view was most ably and logically presented by J. E. Howard,[1] and in the discussion of this paper, as well as in several other preceding publications, Dr. P. H. Dudley, Consulting Engineer of the New York Central Lines, has strongly defended the other opinion.[2] The steel makers of the country have naturally rallied under Mr. Howard's banner, hoping to place the entire blame for these failures upon the railroads with their high wheel-loads, hard steel, heavy traffic, and sometimes inferior road-bed; while the railroad men, on the other hand, have been busy accumulating statistics and records in the effort to find an actual reason in the rails themselves that would account for the seemingly haphazard occurrence of transverse fissures under varying conditions and at widely separated points.

The view that these fissures were due merely to normal fatigue under alternating stresses seemed most reasonable to the writer until quite recently, because from work done in this laboratory, as well as the work of other investigators of the problem, no structural differences, in the vast majority of cases, were found between metal at the nuclei of transverse fissures and metal at similar positions in the same rails or in other rails that had not developed fissures. Within the last year or two, however, since the method of etching polished sections for the microscope with a cupric chloride solution has been tried systematically on lengthwise sections, passing through the nuclei of transverse fissures, evidence

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* Metallographist, Titanium Alloy Manufacturing Co.

[1] *Trans.* (1918) **58**, 597. [2] *Trans.* (1918) **58**, 627.

began to accumulate that there was a certain structural peculiarity of the metal associated with these fissures, very often showing its most distinct development at the nucleus rather than elsewhere in the section examined. So far as the writer is aware, this particular method of examination has not been used by other workers who have published results of metallographic investigations of transverse-fissure rails, so that it may be of interest to present at this time the data that have been acquired through its use.

The cupric chloride reagent that the writer has found to give the clearest and most reliable result in etching polished steel sections for the microscope is the one recommended by Stead.[3] A slightly modified form of this reagent is described in another paper[4] by the same author on the same subject. As used by the writer, it was made by dissolving 2.5 gm. of cupric chloride and 10 gm. of magnesium chloride in 5 c.c. of hydrochloric acid and the smallest possible quantity of hot water, and diluting with alcohol to 250 c.c. Its action on a polished steel surface consists in depositing a film of copper quickly on normal metal, while metal higher in phosphorus remains bright for a longer time. Thus if the action is not continued too long, the location of high-phosphorus areas on a polished section can be shown very clearly as bright spots on a darker background. Silicon, chromium, nickel, copper, and other elements existing in solid solution in the steel, have effects similar to phosphorus under the action of this reagent, but in general it is phosphorus of which the effect is most important and most commonly encountered. The pattern developed by this etching may be made more contrasted by dissolving off the deposited copper with ammonia, as described by Charpy and Bonnerot.[5] This procedure was not followed by the writer in any of the work done on transverse-fissure rails, and the cupric chloride etching was always continued to the same point, as far as could be judged, so that a true comparison in contrasts between different samples was maintained. The effect of the cupric chloride etching on the samples considered in this paper was checked by repolishing and etching again with a 0.5-per cent. solution of picric acid in water, as advised by Stead.[4] This solution darkens the high-phosphorus areas, leaving normal metal brighter; but it was found not so easy to use as the alcoholic cupric chloride, and more apt to act irregularly and show peculiar curved markings that were not duplicated after repolishing and etching again.

Through the courtesy of Dr. P. H. Dudley, of the New York Central

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[3] J. E. Stead: Iron, Carbon, and Phosphorus. *Jnl.* Iron and Steel Inst. (1915) **91**, 140.

[4] Iron, Carbon, and Phosphorus. *Jnl.* Iron and Steel Inst. (1918) **97**, 389.

[5] G. Charpy and S. Bonnerot: Note sur l'Hétérogenéité des Aciers. *Comptes Rendus* (Oct. 22, 1917) **165**, 536–540.

Lines, and a few others, many opportunities have been afforded this laboratory during the past 6 years for investigating the structures of rails that had failed from transverse fissures. At first, when the usual methods of examination were used, such as the making of sulfur prints, examining polished sections for non-metallic inclusions, and sections etched with picric acid or alkaline sodium picrate for the carbon distribution and microstructure, no structural differences were found, in the large majority of cases, between the metal at the nuclei of fissures and that at similar positions in the rail section where fissures had not developed. There were, of course, a few exceptions, where the origin of the fissure could be definitely traced to segregated streaks in the steel, such cases being generally among those in which the transverse fissure had developed from a longitudinal crack. The two sulfur prints shown in Figs. 1 and 2 are illustrations of the exceptional cases where the nuclei of transverse fissures were located directly in segregated streaks that caused dark spots on these prints.

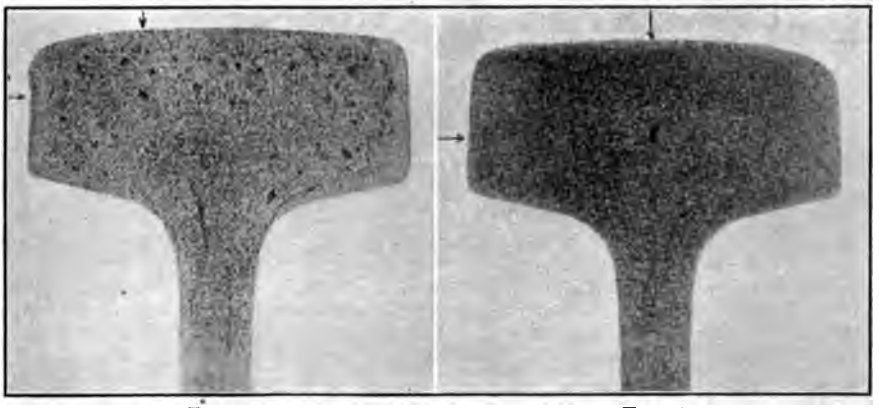

FIG. 1. FIG. 2.

FIGS. 1 AND 2.—SULFUR PRINTS, 7/10 NATURAL SIZE, OF RAILS THAT FAILED FROM TRANSVERSE FISSURES, THE ARROWS POINTING TO THE LOCATION OF THE NUCLEUS IN THE CROSS-SECTION.

When the use of cupric chloride reagents was discovered and advocated by Stead and others for the detection of phosphorus segregation in steels, and this method began to be applied to longitudinal sections cut through the nuclei of transverse fissures in rails, it was found in many cases that the most distinct of the streaks shown in this way passed through the nucleus. In some of these rails the streaks were about the same throughout the section examined, but in practically none were they absent or even indistinct. Fig. 3 is a photograph, taken at a magnification of about 2½ diameters, of two polished microscope specimens etched with the aqueous picric acid solution, which darkens the high-phosphorus streaks. The edges of these specimens show parts of trans-

verse fissures, with the polished sections cutting through the nuclei, and in each case it is plainly seen that the nucleus of the fissure is directly in line with the most distinct dark streak on the polished surface. Figs. 4, 5, 6, and 7 are photomicrographs of the polished surfaces of specimens cut, like those in Fig. 3, through the nuclei of transverse fissures in rails, but etched with cupric chloride. Each of these shows the edge of the specimen at the nucleus of the fissure, and the most distinct high-phosphorus streaks are shown in every case passing directly into these nuclei. Figs. 8 and 9 are similar photomicrographs of other rails that showed longitudinal cracks running through the streaks in line with the transverse fissures. Fig. 9 shows also a transverse crack which was about $\frac{5}{16}$ in from the fissure, and in the same segregated streak as its nucleus.

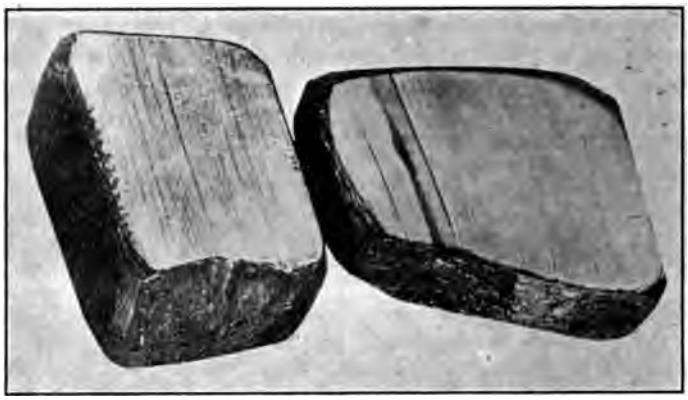

FIG. 3.—TWO POLISHED SPECIMENS FOR THE MICROSCOPE ETCHED WITH AQUEOUS PICRIC ACID AND MAGNIFIED ABOUT 2½ DIAMETERS, SHOWING PARTS OF TRANSVERSE FISSURES ON THEIR EDGES AND THE MOST DISTINCT STREAKS ON THE POLISHED SURFACES IN LINE WITH THE NUCLEI.

Having seen how high-phosphorus streaks were found associated with transverse fissures in all rails that failed in this way and that were examined with reference to such streaks, it would be of interest to look into their causes and determine whether they necessarily exist in all rails, or merely in rails subjected to certain conditions of manufacture. If a definite cause for the streaks can be established, and if some rails can be found without them, it should be easy to show definitely whether or not they really have any connection with the formation of transverse fissures.

The cause of the irregular effect of the cupric chloride and aqueous picric acid solutions on polished steel samples has been demonstrated by Stead, in the papers already mentioned, to be due to the irregular distribution, chiefly, of phosphorus, but partly of other elements in solid solution in the steel. By the well known process of selective freezing which steel ingots undergo after they are poured, the purer metal forms

solid dendrites or "pine-tree" crystals in the liquid metal, which then contains more than its share of the carbon, phosphorus, etc. On cooling further, the dendrites increase steadily in bulk, while the remaining liquid decreases and at the same time becomes more and more saturated with

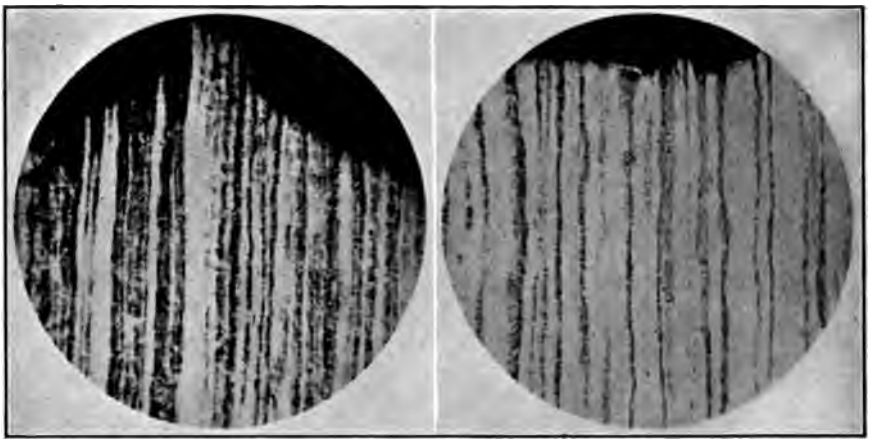

FIG. 4. FIG. 5.

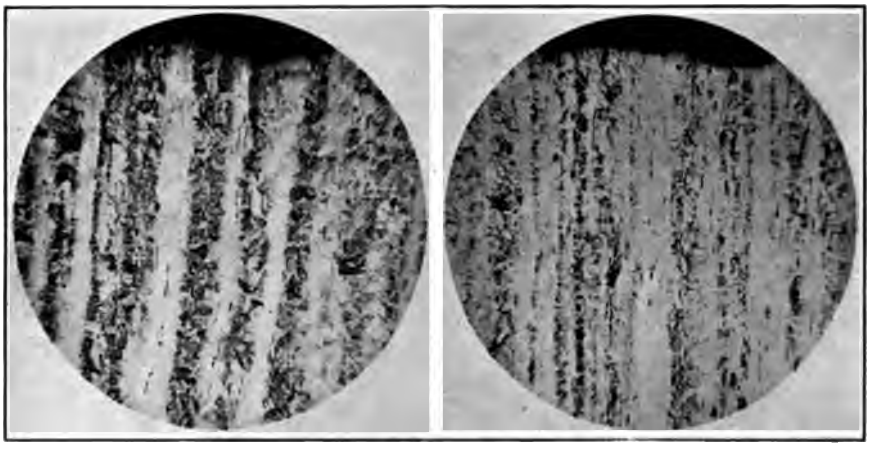

FIG. 6. FIG. 7.

FIGS. 4, 5, 6 AND 7.—PHOTOMICROGRAPHS OF LONGITUDINAL SECTIONS OF RAILS AT THE NUCLEI OF TRANSVERSE FISSURES, ETCHED WITH CUPRIC CHLORIDE AND MAGNIFIED 16 DIAMETERS, SHOWING DISTINCT STREAKS AT THE NUCLEUS IN EVERY CASE.

impurities. Thus the ingot, when first frozen, is far from homogeneous, and can become so only by diffusion in the solid state. This takes place to some extent in the soaking pits, especially in regard to carbon, which becomes thoroughly diffused much sooner than phosphorus, according to Stead and other authorities. When the ingot is rolled out into a rail,

any irregularities in composition are of course rolled into streaks, and instead of a dendritic structure, the result is a lamination. Since the carbon usually becomes thoroughly diffused in the soaking pit, the laminations do not appear in the ferrite-pearlite structure, but since phosphorus diffuses very slowly the laminated structure may often be expected to show in the finished rail when treated in such a way as to reveal the distribution of this element.

Every ingot, when first cast, is chilled quickly on its outside by the cold mold with which it is in contact, and hence its outer skin is nearly homogeneous, and free from coarse dendrites, segregation, etc. For this

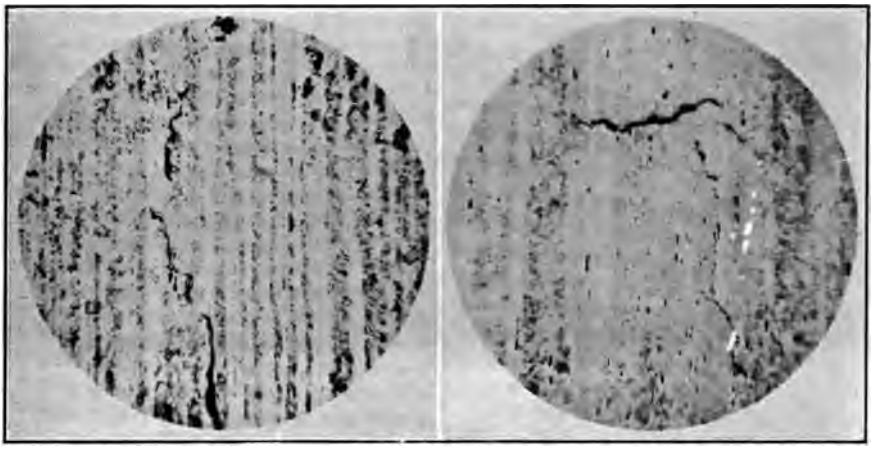

FIG. 8. FIG. 9.

FIG. 8.—A LONGITUDINAL CRACK CONNECTED WITH A TRANSVERSE FISSURE IN A SPECIMEN ETCHED WITH CUPRIC CHLORIDE AND MAGNIFIED 16 DIAMETERS.

FIG. 9.—A LONGITUDINAL AND TRANSVERSE CRACK IN A SEGREGATED STREAK IN WHICH THE NUCLEUS OF A TRANSVERSE FISSURE WAS LOCATED; ETCHED WITH CUPRIC CHLORIDE AND MAGNIFIED 16 DIAMETERS.

reason, the outer surface of a steel rail is always free from high-phosphorus streaks to a greater or less depth. Transverse fissures generally have their nuclei a short distance inward from the surface of the rail section, and it has often been noted that the nucleus will occur at the same distance below the top of the head as the topmost distinct high-phosphorus streak. This streak would, of course, have been subjected to a greater bending moment in service than any streak existing nearer the center of the rail, and hence might be expected to crack first.

The cause of the high-phosphorus streaks has been seen to be selective freezing in the ingot, which cannot be avoided by any means now known. The effects of this selective freezing may, however, be remedied or overcome by diffusion in the solid state, which is a very slow process in regard to phosphorus. If longer heating of the solid steel in rail manufacture will decrease the intensity or distinctness of the high-phosphorus

streaks, by allowing more thorough diffusion, then rails rolled from reheated blooms should show these streaks less distinctly than rails rolled direct from ingots. A most interesting fact in this connection is that out of the few hundred rails that have failed from transverse fissures on the New York Central Lines, just two of them had been rolled from reheated blooms. If it could be shown, therefore, that rails rolled from reheated blooms had the high-phosphorus streaks less in evidence than direct-rolled rails, a strong support would be secured for the theory that these streaks had some influence on the origin of the fissures.

To investigate this point, 24 samples of rails were secured through the coöperation of Doctor Dudley, of the New York Central Lines, 12 of which had failed from transverse fissures, while the rest had given good service in track. All but one of the former had been rolled direct from ingots, and all but two of the latter had been rolled from reheated blooms. Twelve samples of rails that had failed from transverse fissures, that had been previously received from Doctor Dudley, the Louisville & Nashville Railroad, and R. W. Hunt & Co., were also taken into this investigation, making 24 transverse-fissure rails and 12 good service rails. Data received from the senders regarding these samples are shown in Table 1. Unfortunately data as to the amount and character of service endured by the various samples are not at hand, and the method of rolling of the first six samples is not positively known, although there is little doubt that all were rolled direct. These samples represent different heats of steel, with the following exceptions: No. 5 belonged to the same heat as No. 2; No. 7 and 8 were both from another heat; No. 11 and 12 likewise; and also No. 33 and 35. Since the unit for this work seems properly taken as the ingot, with its particular time of heating, it did not appear necessary that each sample should represent a separate furnace charge, even though the results were to be finally averaged for each class of steel. The samples of failed rails represented heats that had produced anywhere from 1 to 10 transverse-fissure rails; No. 21 was one of the two rails rolled from reheated blooms that have been found, on the New York Central Lines, to have failed in this way.

Table 2 gives the results of the metallographic examination of these samples. The first thing done to each was the making of sulfur prints from a cross-section generally within an inch or two of the transverse fissure, provided it appeared on the sample received. These prints were classified as good, fair, poor, or bad, according to the amount of segregation shown. Fig. 1 shows a print from sample 9, classed as "poor," while Fig. 2 shows one from sample 14, classed as "fair." Those classed as "good" showed less segregation than the latter, and those classed as "bad" showed more than the former.

Sections for microscopic examination were cut from the heads of all the samples, longitudinally and upright, that is, parallel to the plane of the

web. The sections covered from ½ to ¾ in. of the height of each head, and were so located as to cut through the nucleus of the fissure in samples that showed transverse fissures. In those that did not, the sections for the microscope were cut near the center of the head. From some of the samples several sections were examined. The grading with respect to presence of alumina, slag inclusions, and distribution of sulfides, was done by the writer from examination of the carefully polished sections before etching, the terms used being descriptive of the quality of the steel

TABLE 1.—*Description of Samples*

Sample No.	Mill	Lb. per yd.	Rail Letter	Method of Rolling	Service in Track
1	A	80	...	Probably direct from ingot.	Failed, transverse fissure.
2	A	90	F	Probably direct from ingot.	Failed, transverse fissure.
3	C	85	...	Probably direct from ingot.	Failed, transverse fissure.
4	A	80	...	Probably direct from ingot.	Failed, transverse fissure.
5	A	90	F	Probably direct from ingot.	Failed, transverse fissure.
6	A	80	...	Probably direct from ingot.	Failed, transverse fissure.
7	B	100	A	Direct from ingot.	Failed, transverse fissure.
8	B	100	A	Direct from ingot.	Failed, transverse fissure.
9	C	105	A	Direct from ingot.	Failed, transverse fissure.
10	C	105	D	Direct from ingot.	Failed, transverse fissure.
11	C	105	A	Direct from ingot.	Failed, transverse fissure.
12	C	105	A	Direct from ingot.	Failed, transverse fissure.
13	C	100	E	Direct from ingot.	Failed, transverse fissure.
14	B	100	A	Direct from ingot.	Failed, transverse fissure.
15	C	100	C	Direct from ingot.	Failed, transverse fissure.
16	B	105	C	Direct from ingot.	Failed, transverse fissure.
17	B	80	C	Direct from ingot.	Failed, transverse fissure.
18	C	100	E	Direct from ingot.	Failed, transverse fissure.
19	B	80	E	Direct from ingot.	Failed, transverse fissure.
20	C	100	B	Direct from ingot.	Failed, transverse fissure.
21	C	105	D	Reheated bloom.	Failed, transverse fissure.
22	B	100	B	Direct from ingot.	Failed, transverse fissure.
23	C	100	B	Direct from ingot.	Failed, transverse fissure.
24	B	80	C	Direct from ingot.	Failed, transverse fissure.
25	D	100	B	Reheated bloom.	Good service.
26	C	105	D	Reheated bloom.	Good service.
27	C	105	B	Reheated bloom.	Good service.
28	E	100	...	Reheated bloom.	Good service.
29	E	100	...	Reheated bloom.	Good service.
30	E	100	...	Reheated bloom.	Good service.
31	B	100	C	Direct from ingot.	Good service.
32	B	100	B	Direct from ingot.	Good service.
33	D	100	...	Reheated bloom.	Good service.
34	C	105	C	Reheated bloom.	Good service.
35	D	100	E	Reheated bloom.	Good service.
36	D	100	C	Reheated bloom.	Good service.

TABLE 2.—*Results of Examination of Samples*

Sample No.	Quality in regard to				Amount of free		Quality in regard to streaks shown by etching with	
	Sulfur print	Presence of alumina	Presence of slag	Distribution of sulfides by microscope	Ferrite	Cementite	Cupric chloride	Aqueous picric acid
1	Good	Poor	Good	Fair	Little	None	Bad	Poor
2	Good	Bad	Good	Fair	Traces	None	Bad	Bad
3	Good	Poor	Fair	Good	None	None	Poor	Bad
4	Fair	Good	Bad	Good	Little	One streak	Bad	Bad
5	Fair	Bad	Good	Fair	Little	None	Bad	Poor
6	Poor	Bad	Good	Poor	Traces	Considerable	Poor	Poor
7	Fair	Good	Good	Fair	Traces	None	Fair	Poor
8	Fair	Good	Good	Poor	None	None	Poor	Poor
9	Poor	Good	Fair	Poor	None	One streak	Bad	Bad
10	Good	Good	Good	Fair	None	None	Poor	Fair
11	Good	Good	Good	Poor	None	None	Bad	Bad
12	Good	Good	Poor	Good	Traces	None	Bad	Bad
13	Good	Poor	Good	Poor	None	None	Bad	Bad
14	Fair	Good	Good	Fair	Traces	None	Bad	Poor
15	Good	Good	Poor	Fair	None	None	Bad	Bad
16	Good	Good	Poor	Fair	Little	None	Bad	Poor
17	Good	Good	Good	Fair	Traces	None	Poor	Poor
18	Good	Poor	Good	Good	Little	None	Bad	Bad
19	Good	Good	Good	Fair	Little	None	Poor	Fair
20	Fair	Good	Good	Good	None	One streak	Fair	Fair
21	Good	Good	Good	Fair	None	None	Good	Good
22	Good	Good	Good	Fair	Traces	None	Bad	Bad
23	Good	Good	Good	Poor	Traces	None	Bad	Bad
24	Good	Poor	Good	Poor	Traces	None	Poor	Poor
25	Good	Good	Good	Fair	Traces	None	Good	Fair
26	Good	Good	Good	Fair	Traces	None	Fair	Fair
27	Good	Good	Good	Fair	Traces	None	Good	Good
28	Poor	Good	Good	Fair	Traces	None	Poor	Fair
29	Bad	Good	Good	Poor	Little	None	Bad	Bad
30	Fair	Good	Good	Poor	Traces	None	Fair	Poor
31	Good	Good	Poor	Poor	Much	None	Bad	Bad
32	Good	Good	Good	Poor	Much	None	Bad	Bad
33	Good	Good	Poor	Good	Little	None	Fair	Good
34	Poor	Poor	Good	Good	Little	None	Good	Good
35	Good	Good	Good	Fair	Traces	None	Poor	Bad
36	Fair	Good	Good	Good	Traces	None	Good	Fair

judged solely on the basis of the particular kind of inclusion under consideration. The data given in the columns headed "Amount of free ferrite" and "Cementite" were obtained by examination of the sections after etching with the usual alcoholic solution of picric acid; and in cases

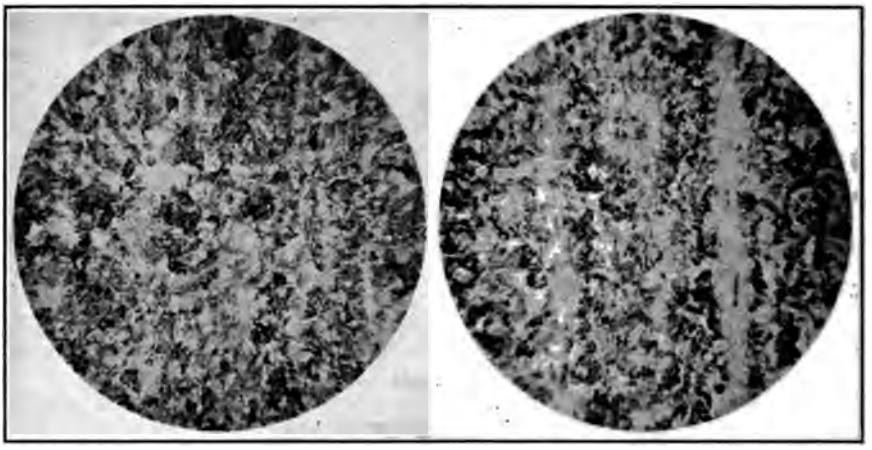

FIG. 10.—GOOD. FIG. 11.—FAIR.

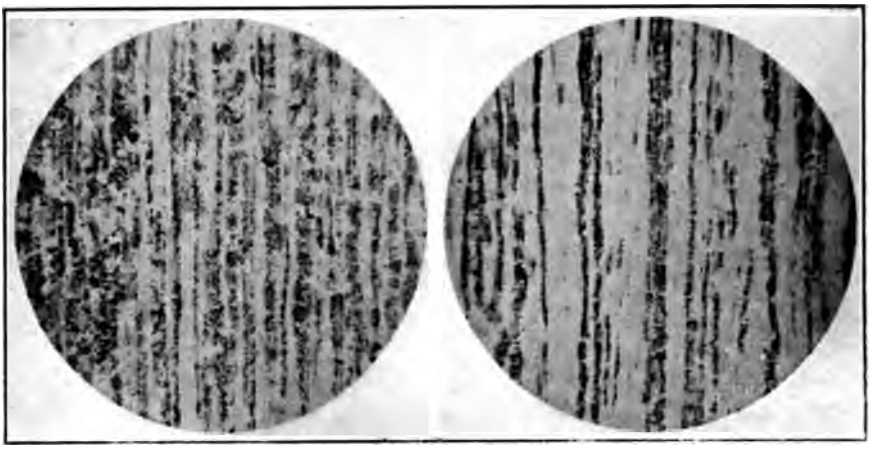

FIG. 12.—POOR. FIG. 13.—BAD.

FIGS. 10, 11, 12 AND 13.—PHOTOMICROGRAPHS OF LONGITUDINAL SECTIONS OF RAIL HEADS ETCHED WITH CUPRIC CHLORIDE AND MAGNIFIED 16 DIAMETERS, TO ILLUSTRATE THE SIGNIFICANCE OF THE TERMS USED TO DESCRIBE THE QUALITY OF SPECIMENS WITH REGARD TO HIGH-PHOSPHORUS STREAKS, SHOWN BRIGHT BY THIS ETCHING.

where cementite was present this was checked by repolishing and etching with boiling alkaline sodium picrate. The term "one streak" in this connection does not mean that only one particle of free cementite was seen in the section, but that there was in the sample one segregated

streak containing an appreciable network of free cementite between the pearlite grains.

The samples were next repolished and etched with Stead's cupric chloride reagent, all as nearly as possible to the same degree, and were

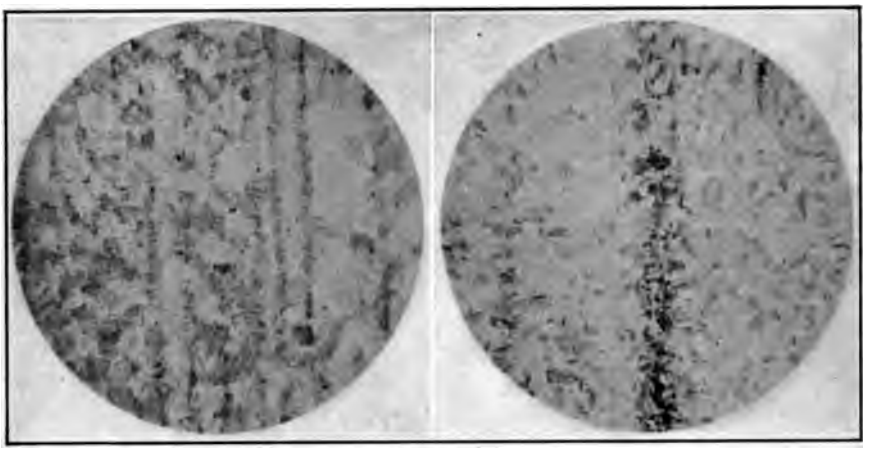

FIG. 14.—GOOD. FIG. 15.—FAIR.

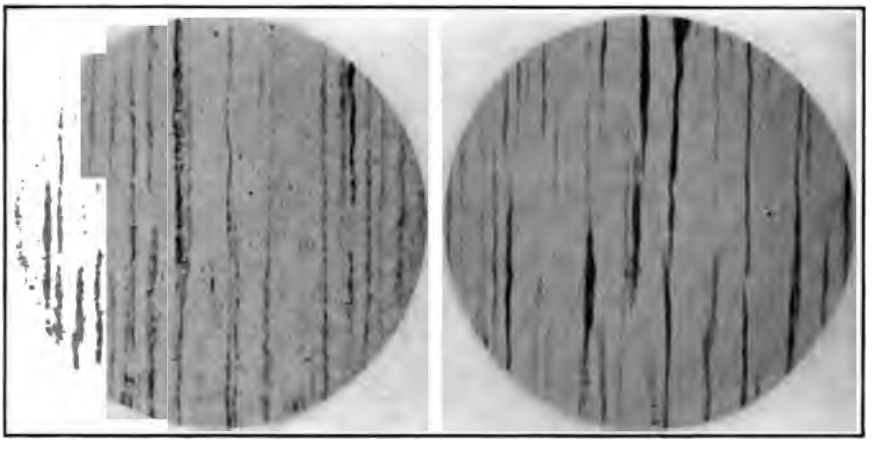

FIG. 16.—POOR. FIG. 17.—BAD.

FIGS. 14, 15, 16 AND 17.—PHOTOMICROGRAPHS OF LONGITUDINAL SECTIONS OF RAIL HEADS ETCHED WITH AQUEOUS PICRIC ACID AND MAGNIFIED 16 DIAMETERS, TO ILLUSTRATE THE SIGNIFICANCE OF THE TERMS USED TO DESCRIBE THE QUALITY OF SPECIMENS WITH REGARD TO HIGH-PHOSPHORUS STREAKS, SHOWN DARK BY THIS ETCHING.

classified in the same way as the sulfur prints, according to the quality of the metal judged from the distinctness of the streaks shown. Figs. 10, 11, 12, and 13 illustrate the terms used in describing the quality after this etching. Finally the samples were all repolished again and etched

for 15 to 20 sec. with the aqueous picric acid solution recommended by Stead, and graded as before. Figs. 14, 15, 16 and 17 illustrate the significance of the terms for this grading.

TABLE 3.—*Classification of Samples*

In regard to	Grade	Transverse-fissure rails, per cent.	Rails that gave good service, per cent.	Rolled direct from ingot, per cent.	Rolled from reheated blooms, per cent.
Sulfur print	Good	67	58		
	Fair	25	17		
	Poor	8	17		
	Bad	..	8		
Presence of alumina	Good	67	92		
	Poor	21	8		
	Bad	12			
Presence of slag	Good	75	83		
	Fair	8			
	Poor	13	7		
	Bad	4			
Distribution of sulfides by microscope	Good	21	25		
	Fair	50	42		
	Poor	29	33		
	None	38			
Amount of free ferrite	Traces	37	58		
	Little	25	25		
	Much	..	17		
	None	83	100		
Amount of free cementite	One streak	13			
	Considerable	4			
Cupric chloride etching	Good	4	33	..	46
	Fair	8	25	11	27
	Poor	29	17	26	18
	Bad	59	25	63	9
Aqueous picric acid etching	Good	4	25	..	36
	Fair	12	33	16	36
	Poor	38	9	32	10
	Bad	46	33	52	18

In Table 3 these results are all averaged for the rails that failed from transverse fissures and for those that endured good service without failure, in order to get the average classification of each kind of rails with respect to the different characteristics that were examined. For instance, in regard to sulfur prints, out of 24 transverse-fissure rails, 16 were classed as "good" in this respect, or 67 per cent.; 6 were classed as "fair," or 25 per cent.; and the remainder, or 8 per cent., as "poor." The classification of direct-rolled rails was practically the same as that for the transverse-fissure rails, and the rails rolled from reheated blooms were also

similar to the good-service rails, so that it did not seem desirable to fill out the last two columns of the table except in regard to the high-phosphorus streaks.

In Table 3 it is readily seen that the majority of both the failed and the good rails were classed as "good" in regard to sulfur prints, presence of alumina, and presence of slag; and as "fair" in regard to distribution of sulfides, with either no ferrite or only traces, and with practically no cementite. Of course there are slight differences in these respects, and the failed rails are seen to average better in regard to sulfur prints, worse in regard to alumina and slag inclusions, and to have, on the average, less ferrite and more cementite than the good rails. The showing in regard to sulfur prints should not be taken to mean that rails showing segregation in these prints are more apt to give good service, without failure, than rails of which the sulfur prints are good, for this investigation is concerned with only one particular type of rail failure, and there is no question whatever about segregated rails being especially susceptible to another type of failure known as a "split head." These results show, however, that segregated sulfides are not a cause of transverse fissures. In regard to alumina inclusions, the results might be considered as indicating that these have some effect on the fissures were it not for the fact that no case has arisen where a streak of alumina particles was found in line with and close to the nucleus of a fissure. The showing in regard to free ferrite and cementite is merely a corroboration of the fact several times stated by other writers that transverse fissures occur most frequently in hard high-carbon rails. The indication is worth noting, however, that a thick ferrite network in the microstructure seems to mean less tendency toward the formation of fissures, while cementite-bearing streaks in the head mean a greater tendency toward their development.

The various aspects of the structure so far considered in Table 3 have given, for the two kinds of rails, classifications that were similar in their general indications, showing that none of these peculiarities could be an important cause of transverse fissures. The last two criteria, however, show a decided difference in the classification figures for the two kinds of rails, and indicate that there is here at least some relation between these aspects of the structure and the formation of the fissures. Thus, while 88 per cent. of the failed rails were classed as poor or bad in regard to the high-phosphorus streaks shown by the cupric chloride etching, only 42 per cent. of the good-service rails were so classed; and the aqueous picric acid etching checks these figures in almost the same way. Comparing the direct-rolled rails with the rails rolled from reheated blooms, the figures are more strongly suggestive, becoming 89 and 88 per cent., respectively, for the sum of the poor and bad classes of the direct-rolled rails after the two methods of etching, and only 27 and 28 per cent. for

the same classes of the rails rolled from reheated blooms. In obtaining these average classification figures for direct-rolled rails, samples 1 to 6, inclusive, had to be omitted on account of the slight uncertainty regarding their method of rolling. But on the 30 samples of which the method of rolling is known without doubt, the cupric chloride etching shows regularly increasing percentages from the "good" to the "bad" classes for the direct-rolled rails, while the percentages regularly decrease from "good" to "bad" for the rails rolled from reheated blooms. The aqueous picric acid etching checks these figures fairly closely. Here is, then, a clear indication that reheating the blooms in rail manufacture will give a product decidedly more free from high-phosphorus streaks than the direct-rolling process will give, and it has also been shown that rails can be made, and are made, that are practically free from these streaks (see Figs. 10 and 14).

The inclusion of samples 1 to 6 among the direct-rolled rails, where they belong in all probability, would alter the figures in the third column of Table 3 to only a slight extent. The difference between the classification in the first two columns of this table and that in the last two columns is due, as shown in Table 1, largely to three samples—No. 21, 31, and 32. The former, as stated before, is one of the two rails that were rolled from reheated blooms that have failed from transverse fissures on the New York Central Lines, and hence is an exceptional case. The other two, both from the same mill, were rolled direct from the ingots, and showed very distinct high-phosphorus streaks, yet gave good service without developing fissures. As seen in Table 2, these two rails are the only ones in the entire series that showed much ferrite in their microstructures, and indeed they had every appearance of being Bessemer steel, although reported as open-hearth. It may be, therefore, that the presence of the thick ferrite network helped these rails to resist the development of fissures in spite of their high-phosphorus streaks, and this may likewise be the reason why Bessemer rails, in general, do not develop transverse fissures in spite of the streaks which would be expected from their higher phosphorus content. If these two rails were considered as exceptional, on account of being lower in carbon than all the other rails, and were discarded from the averages, together with the exceptional No. 21, the classification figures in Table 3 for the rails that gave good service would become practically identical with those for the rails rolled from reheated blooms, which offer a more striking comparison with the figures for the failed rails.

Attention should also be drawn to the three samples 28, 29 and 30, which were the only rails rolled by mill "E," and were received from a certain railroad where the service may not have been as severe as that which some of the other rails endured. These rails were the worst of all in regard to sulfur prints, and also made a poor showing in regard to high-phosphorus streaks, although reported as having been rolled

from reheated blooms. There is thus a suspicion that the mill practice was not the best where these rails were made, and that their service record might not have been so good had they been on the same lines as the other samples here considered. These rails were included among the averages in the "good service" and "reheated bloom" columns of Table 3, but had they been omitted on account of being from a different mill and railroad from the other samples, the showing in these columns would have been better.

Summarizing the evidence connecting transverse fissures with the high-phosphorus streaks shown by etching longitudinal sections of the rails with a cupric-chloride reagent, it has been shown that in many instances the nucleus of the fissure is located in the most distinct of the streaks in the rail head, or in the uppermost of the distinct streaks, and that practically all rails with this type of failure that have been examined have been found to contain these streaks. Also, it has been shown that of the rails examined, those rolled from reheated blooms were much less apt to contain these streaks, due to uneven phosphorus distribution, than the direct-rolled rails were, and many that gave good service in track, without failure, lacked the streaks entirely. This fact fits in very well with the discovery of Dr. P. H. Dudley, of the New York Central Lines, that rails rolled from reheated blooms do not, with two exceptions to date, develop transverse fissures, and the natural inference is that it is because they do not contain the high-phosphorus streaks. This gives a definite reason both from practice and theory for reheating the blooms in rail manufacture.

It is interesting to note also in this connection that W. R. Shimer[6] concluded that rails rolled from reheated blooms are more ductile than direct-rolled rails, regardless of their finishing temperature. He explained this by referring to the removal of rolling strains in the reheating furnace, an explanation which was not considered valid by Professor Hoyt in discussing the paper. Probably a better diffusion of phosphorus was the chief reason for the better ductility found by Mr. Shimer in his rails rolled from reheated blooms, and this aspect of the microstructure was not mentioned either in his paper or in the discussion.

It should be understood that the high-phosphorus streaks need not be always present in direct-rolled rails, because if the ingot was kept in the soaking pit long enough the phosphorus would diffuse there just as well as in the reheating furnace for blooms, although it would probably take longer for thorough diffusion to occur in an ingot than in a bloom. Furthermore, a rail rolled from a reheated bloom might readily show distinct streaks if both the ingot and bloom had been heated only as short

[6] Effect of Finishing Temperature of Rails on their Physical Properties and Microstructure. *Trans.* (1915) **51**, 828.

a time as possible. Nevertheless, the chances for a direct-rolled rail to show a distinctly streaked condition in regard to phosphorus are much greater than in the case of rails rolled from reheated blooms, as is shown by this investigation.

It is not claimed that reheating the blooms is a sure cure and the only cure for transverse fissures in rails, because the presence of a fissure in sample 21 would refute this claim. But the evidence here given supports strongly the contention that segregation of phosphorus in abrupt alternate bands of almost microscopic size, running lengthwise in the rail-head, are an important cause of transverse fissures, and that by reheating the blooms from which rails are rolled these bands may be reduced in intensity by diffusion and the tendency toward the formation of transverse fissures may be appreciably lessened.

These conclusions do not contradict or disprove the fatigue-failure theory of transverse fissures, although they seem to support the opposite view of origin from a defect in the steel. Mr. Howard's work has been valuable in showing how the peculiar appearance of these fissures is caused and in giving a reason for their failure to start at the upper surface of the rail. But may not these views of his be accepted, without denying the view of the other school that there must be some reason in the steel for the appearance of these failures in much greater numbers in rails from certain heats, or in rails from certain periods of rolling, than in other similar rails subjected to exactly the same service? The point missed by the supporters of the fatigue-failure theory is that even a fatigue failure must have a starting point, and that it is very reasonable to suppose that all possible variations in structure would not have exactly the same influence on the facility with which such a fatigue failure would start. It is claimed that transverse fissures can be produced mechanically at will at any point desired in any rail, which is interesting as showing that the mechanism of growth of these failures is now well understood. But almost any type of failure can also be produced at will, mechanically, at any definite point, and the important factor is to know what internal conditions in the material tested make the failure start more easily, and what conditions make it start with more difficulty or only after a longer time. This paper is presented with the idea of throwing a little light in a new direction on the internal conditions that help the beginning of transverse fissures in rails, and in the hope that it will be of some use in linking together the two opposing views as to the origin and growth of these peculiar failures.

DISCUSSION

JAMES E. HOWARD,* Washington, D. C. (written discussion†).—Mr. Comstock, in his classification of the possible causes of transverse fissures, has not quite caught the point of view of the writer. It appears,

* Engineer-physicist, Interstate Commerce Commission. † Received Jan. 18, 1919.

following the trend of thought in Mr. Comstock's paper, that three groups instead of two might be enumerated, namely: Fatigue of the steel, quality or, preferably, grade of steel, and mill practice.

All steels are liable to failure by fatigue. Stresses that are capable of causing fatigue fractures are modified in their magnitude by the grade of steel acted upon. Mill practice is a term having a very wide meaning and in its relations to the development of transverse fissures has been used with such vagueness as to include many known factors and some that are unknown in the arts. According to the grade of steel employed, fatigue fractures may be accomplished by loads of greater or less magnitude, hence there is a relation between quality, or grade, of steel and the display of fatigue fractures. Mill practice of such an order as to affect the formation of transverse fissures must be shown to exert an influence on the ability of the steel to endure repeated stresses. Different grades of steel are selected for different purposes. Steel employed for fire-box purposes is of different grade from that used for cutting tools, although each grade is suitable for its purpose, and therefore of good quality.

It is conceivable that a demand might be made for steel of physical properties in excess of those of any known grade. Under such a demand, impossible of fulfillment, it would not be proper to refer to the steels as being deficient in quality; to do so would be a failure to recognize that all steels have their limitations in strength. Service conditions peculiar to rails admit of successive advance until the conditions can no longer be met. The failure of a large number of rails from one type of failure, when all known means of averting the failure have been employed without success, is sufficient evidence that the end is close at hand.

Iron rails, for a time, successfully met the requirements of railway service but weight of equipment, motive power, and rolling stock were increased, necessitating the use of steel. The early steel rails enjoyed the reputation of being of excellent quality, although neither chemically nor structurally would they be so regarded at the present time. They did good service in their day and under the conditions then prevailing were good rails. To meet the increased weight of rolling stock there has been an increase in the weight of rails, at the same time practically all grades of carbon steel have been tried, with occasional efforts to use alloy steels and heat-treated rails. No grade of steel nor weight of rail yet employed has enjoyed immunity from fracture in service.

In the first report dealing with a rail fracture to which the term transverse fissures was applied, the writer expressed the opinion that this type of fracture was "due to the use of a hard steel and subjecting it to high wheel pressures, . . . causing the development and introduction of internal strains, and those internal strains together with the wheel loads caused the development of interior fissures and their extension to a dangerous degree."

It will be noted that attention was called to the presence of internal strains as a result attending the use of rails in the track. The presence of internal strains, which at times far exceed the direct stresses, had apparently been entirely overlooked on the part of railroad engineers, an omission to this day not completely overcome. Internal strains are introduced in the upper part of the head of the rail, where a range in stress from tension to compression of 28,000 lb. per sq. in. (19.68 kg. per sq. mm.) has been reached, within a depth of metal not more than ¾ in. (19 mm.). Compressive stresses located next to the running surface of the head not infrequently exceed 20,000 lb. per sq. in. (14 kg. per sq. mm.). With the occurrence of thousands of transverse fissures in recent years—they were known only in isolated cases in earlier years—the query is forcibly presented whether service conditions have not reached that stage in which the ability of steel to endure them has nearly or quite reached its limit.

Increased weights of rails furnish greater girder strength, without material modification of the impinging pressures between the wheel and the rail. The interior origin of a transverse fissure is due to the internal strains introduced by these impinging pressures. Increased weight of rail does not remove this compressive component in the head. Transverse fissures occurring in the heavier rails after a short period of time in the track, in rails in which the direct bending stresses are much reduced, place a large share of responsibility for failure on the cold-rolling action of the wheels.

Interest centers upon the condition of the steel at the nucleus of the transverse fissure, whether the inception of the fissure was caused by any local abnormality or was due entirely to repetition of stresses. No structural or chemical reason common to the display of transverse fissures has been found to which their formation might be ascribed. In given heats, these fissures have been found in rails that were laid on divisions where traffic conditions were most severe. Their occurrence in hard rails is probably facilitated by reason of the heads of the rails maintaining their shape with minimum distortion. After reaching a saturated state of internal strain, the maximum stresses thereafter are centered upon the same elements, a favorable condition for the development of fatigue fractures.

The nuclei of transverse fissures commonly present the silky appearance so much admired in steels fractured by tension. The surrounding parts have a bright silvery luster, until the fissure has extended and reached the periphery of the rail; then the surfaces are darkened. The silvery luster is doubtless acquired by the opposite surfaces hammering each other, when the stresses in the rail are reversed and the head is in compression. The progressive development of transverse fissures is shown by the concentric ripples on their silvery faces.

The walls of an incipient fissure are separated a very minute distance, probably represented by the elastic resilience of the steel in that vicinity. The phenomenon of permanent set or permanent elongation of the steel cannot take place when an interior fissure is formed in a matrix of structurally sound metal. Seamy or spongy steel would appear to favor interior flow of limited degree without fracture.

Let us consider for a moment some features in the physics of steel that must have a bearing on the inception of a transverse fissure. To be a potent influence in starting a transverse fissure in the interior of the head, in the absence of those internal strains that we know are introduced by the wheel pressures, there must be a portion of the metal devoid of such elastic properties as will prevent its extension and compression in harmony with the surrounding metal. It is not a question of toughness or brittleness in the usual sense of those terms, but one of elasticity. We are not familiar with any alloy of iron that does not possess elastic properties; steels covering a wide range in chemical composition still retain the same, or substantially the same, modulus of elasticity. Hook's law prevails also in steels prior to the application of overstraining loads. Any minute eddy in the transmission of stresses will be a negligible factor since the effect of such an eddy will merely be a moderate increase in intensity of stress in that vicinity. A non-elastic substance of considerable volume and having sharp edges will be required to cause the incipient formation of a transverse fissure. The reduction of the metal from the ingot to the rail orients seaminess and lamination parallel to the length of the rail. Interruption of continuity normal to the principal stresses is necessary in order to influence rupture.

The strains per unit of length with which we are concerned in connection with transverse fissures are less than two thousandths. Such strains in the field of the microscope represent very minute distances—millionths of an inch. It would be difficult to recognize their presence even if their characteristics, as means of identification, were known. Recalling Tyndall's familiar illustration in molecular physics, comparing an orange with the size of the earth, it seems doubtful whether the microscope brings us relatively much nearer a real conception of the phenomenon of fatigue in metals than the unaided eye.

Physical properties that are obviously correlated with stresses that ultimately result in fatigue fractures have been taken under consideration in the study of transverse fissures. It is well known, (1) that the moduli of elasticity of steels are lowered by overstraining loads and that a recovery takes place during a period of rest; (2) that overstraining loads disturb the equality of elastic limits in hot-worked steel, overstraining in one direction and impairing the elastic limit in the opposite direction; (3) that internal strains are acquired by steels when cold worked and also during periods of cooling; (4) that hot-forged steel shows a difference in behavior,

when annealed, in respect to the direction the metal was worked and at right angles thereto. It is also known that, with changes in carbon content, there are changes in the coefficients of expansion and in the specific-gravity values of steels. Each of these factors may have an influence on the ability of a rail to endure stresses, either from their influence on the primitive state of the rail or by affecting its condition in the track. The relations between the properties made use of in rails in service and their microconstituents are less direct and less understood.

A slight initial advantage against the acquisition of internal strains would seem to attend segregated steel, a thought at variance with specifications in which higher carbon metal in the center of the head is a cause for rejection of the rail. The lower carbon steel has the higher coefficient of expansion, in which case a nucleus of high-carbon metal in the head should acquire a state of compression. The higher carbon steel also has a lower specific gravity, and if an equal weight of the high-carbon metal was forced into the space that the same weight of low-carbon steel would occupy there would result cubic compression of the segregated metal. What happens along these lines we do not know, whether favorable or unfavorable in resisting fatigue. So great is the influence of the rate of cooling on the initial strains that the influence of other factors is overshadowed.

Sponginess of structure is occasionally witnessed in steel in the early passes in the blooming mill, disappearing commonly before the finishing pass of the rail mill is reached. Etched specimens from rail-heads have exhibited similar manifestations. The presence of such indications is of interest and raises the question whether the interior sponginess was not advantageous in resisting the formation of transverse fissures by diffusion of the internal strains. The most defective steel in fissured rails has been found in those rails that resisted the formation of transverse fissures the longest intervals of time in the track. So far as this evidence goes, it does not increase our anxiety as to the detrimental effect of seaminess as a factor in leading to the formation of transverse fissures.

Concerning the influence of longitudinal seams, streaks, or laminations upon the development of transverse fissures, it is not known that they exert any direct influence. The iron-phosphide streaks, forming the subject matter of the paper under discussion, are oriented in the same manner as other streaks; they are parallel to the length of the rail, in which direction the influence of streaks is not materially felt on the tensile properties.

The American Railway Engineering Association has published one or two reports upon interior fissures, in which longitudinal seams were referred to. The distinction will be kept in mind between such manifestations, which are commonly called seams, and transverse fissures, the

similarity in captions tending toward confusion. Puddled iron may be referred to as an example of seamy fibrous metal, oriented longitudinally, upon which experiments have been conducted on repeated stresses. Wrought-iron bars, in terms of the elastic limits, have shown results superior to those of high-carbon and medium-carbon steels. A wrought-iron bar endured a fiber stress 36 per cent. in excess of its elastic limit some eight times longer than a medium-carbon steel endured a fiber stress less than 2 per cent. in excess of its elastic limit.

In the minds of some there are features associated with rails that are regarded as causes of the formation of transverse fissures, which upon analysis appear as deterrents. To the writer, it seems preferable to assign as causes of the formation of transverse fissures those factors that are known to lead to rupture rather than to pin faith upon features that are still obscure and the correlation of which with transverse fissures has not yet been established.

P. H. DUDLEY,* New York, N. Y. (written discussion†).—The author limits his paper to this cause of interior transverse fissures in steel-rail heads. This renders his contribution important as to a cause and its diffusion as a remedy, for as I now recall the outcome of preceding work, it is the only one discovered by metallographic research in the many years of investigation of the subject. The remedy assigned of reheating the blooms is old, and originated with the manufacture of Bessemer steel rails, therefore it is fully established by mill practice and confirmed by years of service tests in the railroad tracks. When the mills for the manufacture of Bessemer steel rails were first installed in the United States, the blooms were reheated before the rails were rolled. John Fritz and Alexander Holly always constructed furnaces for reheating the blooms in the mills. They knew by intuition and experience that the steel rails would have greater ductility to withstand breakages in service. They tested the steel of each Bessemer heat by the simple method of bending, to fracture around the horn of an anvil, a forged or rolled bar $\frac{3}{8}$ by $\frac{5}{8}$ in., from the test ingot. The texture of the fracture of the steel bar was examined with the same scrutiny and care as crucible tool steel, for at the inception of the manufacture of Bessemer steel rails, they did not have chemists to make an analysis of the steel. The first manufacturers, therefore, were obliged to judge the requisite properties of the steel for rails by the bending tests, fractures of the small bars, and small test ingots. Drop tests on the rails were installed in a few years after the construction of the mills. The organization of this Institute, nearly half a century ago, had for one of its objects the discussion of Bessemer steel-rail manufacture, and then to make the records available to a greater number of interested railroad officials.

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† Received Feb. 17, 1919.

The iron rails, because of the 1 or 2 per cent. of included slag, failed so rapidly when the driving wheel loads had increased to 10,000 or 12,000 lb. on the Eastern railroads prior to 1860, that the tracks could not be maintained, regardless of cost, in as good condition as desired. The railroad officials were anxious and desired Bessemer steel rails to replace those of iron. I remember a railroad president saying "iron rails were but a bundle of fibers while Bessemer steel was melted and homogeneous." Bessemer steel rails were made low in carbon and phosphorus in acid-lined converters in large quantity. This enabled great systems of railroads to be constructed to provide a new country with ways and communications for enterprising peoples with faith in their ability and vision of future achievement.

All of the early Bessemer steel-rail manufacturers retained the practice of reheating the blooms. One Bessemer mill was constructed, when the output of this grade of steel was less than the demand, without this furnace, in order to reduce the time and cost of manufacture. The reheating furnace was also omitted in some subsequent mills, but these reheating furnaces would not have been omitted in the designs of the plants had the constructing officials understood that the quality of the output would be affected in an occasional rail-head. Their value is now being recognized and reheating furnaces will doubtless be installed, in a short time, in all basic open-hearth mills that did not include them in their original construction.

I am pleased to contribute statistics that support the paper of Mr. Comstock that the mill practice of reheated blooms, as a process of manufacture, does reduce interior transverse fissures in the rail-heads of the output compared to the output of rails rolled direct from the ingot under their own initial equalized heat. I noticed from the collection of statistics of service failures that rails from reheated blooms developed hardly any interior transverse fissures while they were more abundant in rails that were rolled direct. The service records of eight heavy-traffic railroads show that of 882,237 tons of rails that failed from interior transverse fissures 559,644 tons were rolled direct, developing 1054 interior transverse fissures. Of 322,593 tons from reheated blooms, only 59 developed interior transverse fissures and nearly one-half of these failures were from rails rolled when the mill was being equipped for reheating the blooms. Some of the other failures were from rollings of rails when reheating of the blooms was under difficulties not common to the regular mill practice for such work. These rails were rolled at eleven different mills in and between the years 1909 and 1915, while the record of failures includes those that developed up to Oct. 1, 1918.

One mill that reheats the blooms has produced, in the past 12 years or more, a large output of basic open-hearth rails in which up to Jan. 1, 1919, I have not been able to find, from the different railroad companies

using the rails, that a single interior transverse fissure has occurred in the rail-heads. Another mill, with not as large an output, has the record of only one interior transverse fissure.

The cause described by Mr. Comstock is a new announcement, but there are other causes of interior transverse fissures that have been identified and studied for nearly four years, which should be included in this discussion. See also my Report No. 68 to Rail Committee, A. R. E. A., Dec., 1917.

Cause and Effect.—My investigations show that interior transverse fissures in rail-heads are induced by a combination of two or more exceptional conditions of manufacture in an occasional rail-head principally by direct rolling, which can be practically eliminated by reheating the blooms. Induced interior transverse fissures in basic open-hearth rails are due, in part, to an occasional hot bar being cooled so rapidly by the rolls and their pressure to elongate the rail section or then chilled on the hot beds by gusts of air before recalescence as to cause a lag of some of the transformations of the metal in the interior of the head. These fissures can only develop in the track from the effect of the preceding causes which are known and their progressive stages of development occur in the following order:

The first contributory cause may be the segregation of the setting ingots, delayed transformations of the steel near the center of the head, chemical and mechanical defects. The first effect is induced loss of strength, initial stresses, a non-ductile core of physically heterogeneous metal in the interior of the head, with an occasional zone of enlarged grain growth.

The second contributory cause may be mechanical injury to the physically heterogeneous metal. Either the nucleus in the rail-head is checked transversely by the gag, when it is applied on the base of a low rail or the imprint of the gag is checked longitudinally when applied on the head of a high rail. See descriptions of Figs. 18 and 19. The second effect is the produced defect of an interior invisible part of injured and checked area of metal in the interior of the head before the rail leaves the steel plant.

The third contributory cause may be that the rail section functions as a girder in the track to carry and distribute the rolling wheel loads; the stresses directly under the wheels in the rail-head are compressive, while in the base underneath the neutral axis they are tensile. The tensile stresses in the rail-head occur only in the wheel spacing between the points of flexure for each passing wheel load, and are about one-third or less in magnitude of the tensile stresses in the base directly under the wheels. Compressive stresses in the rail section are simultaneously developed opposite the tensile stresses in either the head or the base, and each is alternately in tension or compression to carry and distribute the rolling

wheel loads. It is in this manner that any metal in the interior of the rail-head, when in the wheel spacing, is under the longitudinal stresses of tension, and a broken area is susceptible of enlargement.

The third effect is that the broken area of interior metal in either type is no longer able to restrict either the pressure or the stresses of the wheel

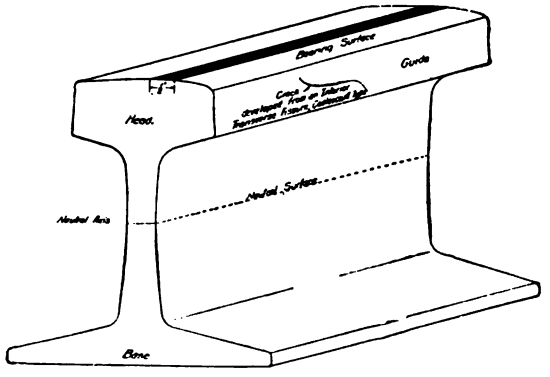


FIG. 18.—A 6-IN. 100-LB. SECTION THAT SHOWS A CRACK ON SIDE OF HEAD WHICH IS COMMON TO COALESCENT TYPE OF FISSURE, DUE TO UPSETTING OF METAL TO SHORTEN HEAD AND LENGTHEN BASE.

loads, as the case may be, within the usual range of elastic limits of the steel, as in sound metal, which is homogeneous. Then, due to the injury in manufacture, there develops by "detail" growth the specular surfaces from and around the circumference of the broken areas of the

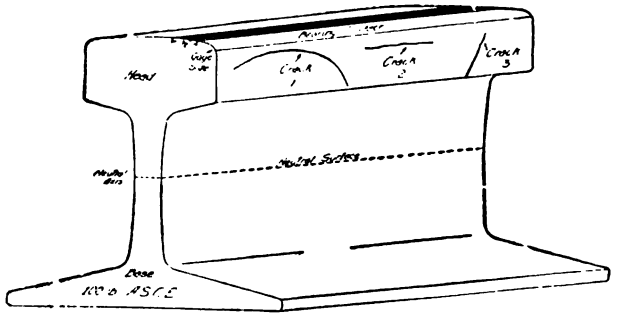


FIG. 19.—A 100-LB. A. S. C. E. SECTION. CRACK 1 SHOWS A FORM FOUND IN A FEW RAILS; CRACK 2 IS GENERAL FOR COALESCENT TYPE OF FISSURE, AND MAY BE FROM 2 IN. TO 3 FT. LONG. CRACK 3 SHOWS THAT NEARLY VERTICAL INTERIOR TRANSVERSE FISSURE HAS WORKED OUT TO SURFACE AND HAS BEEN FOUND IN SEVERAL RAIL-HEADS. COALESCENT TYPE OF FISSURE, WHEN UNCAPPED, SHOWS IMPRINT OF GAG IN ALL CASES, AS IN FIG. 22.

induced interior transverse fissures until the section generally ruptures. When the induced interior transverse fissures by growth crack through the metal to the air, the specular surfaces discolor and darken, as found in ruptured rails by the trackmen.

The service records of Bessemer or basic open-hearth steel rails show that interior transverse fissures have not occurred in the countless thousands of rail-heads of physically homogeneous metal under the same wheel loads. The reason for their non-occurrence is that the rail-heads did not contain the conditions of the first contributory cause, consequently, while the metal is lengthened or shortened by the gag it was not checked. Three interior transverse fissures only have developed to date in basic open-hearth rails from the chemical composition prescribed by the New York Central Lines' specifications and rolled from reheated blooms.

Classification and Illustrations.—The specimens in my collection after investigation, I classified by June 9, 1915, into two general types: intergranular and coalescent.

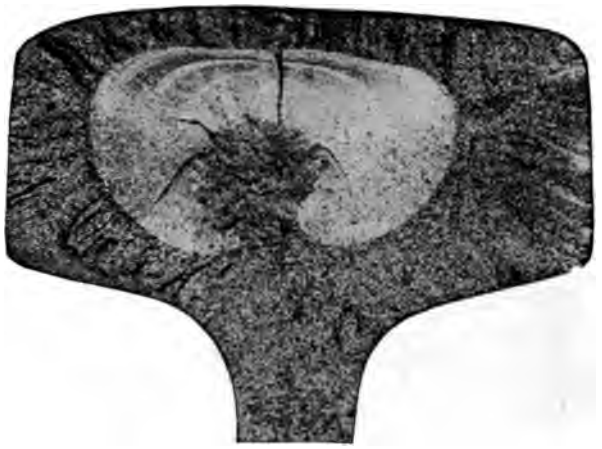


FIG. 20.—INDUCED INTERIOR TRANSVERSE FISSURE OF INTERGRANULAR TYPE. NUCLEUS IS 0.625 IN. IN DIAMETER, THOUGH SMALLER IN A MAJORITY OF SPECIMENS.

In the intergranular, the nucleus, in effect, physically is a non-ductile core in the head which is checked between the grains of the metal by the gag of the straightening press, and may be 0.625 in. in diameter, or as small as 0.0625. The subsequent development in the track of this type of interior transverse fissure is from the checked circumference of the area of the nucleus, which progresses through the grains by a detail growth of specular surfaces, in striking contrast to the check between the grains, the effect of the blow of the gag. This type develops in rails, or part of a rail, which cooled low on the hot beds and then were gagged upon the base to shorten the metal and lengthen that of the head. The nucleus in Fig. 20 is over 0.625 in. in diameter, though smaller in a majority of the specimens.

Three characteristic fractures of the metal in the final rupture of the section occur in the induced interior transverse fissure of the intergranular type: The nucleus checked between the grains of the metal by

FIG 21.—COALESCENT TYPE, SHOWN FULL SIZE, 100-LB. A. S. C. E. RAIL. THE CRACK ENLARGED BY BEING UNCAPPED FOR INVESTIGATION, AS SHOWN IN FIG. 22.

FIG. 22.—COALESCENT TYPE, FULL SIZE, SHOWS IMPRINT OF GAG, ALSO CAP. THERE IS ALSO SHOWN "DETAIL" GROWTH BY WHEEL LOADS FROM CHECKED LONGITUDINAL *NUCLEUS*.

the gag; the "detail" growth through the grains of the specular surfaces; the final fracture of rupture.

In the coalescent type, there is a checked longitudinal elliptic nucleus, which develops from the imprint of the gag on the non-ductile metal, 0.5 to 0.625 in. under the bearing surface of the head, and enlarges to a longitudinal fissure by the rolling wheel loads. This fissure coalesces through a transverse or oblique check into the nearly vertical interior transverse fissure, which usually develops to rupture after two or more years of service in the track. The fracture is often conchoidal at the junction of the coalescence on the side affected by the maximum shearing stresses produced by the gag.

FIG. 23.—COALESCENT TYPE, FULL SIZE OF BEARING SURFACE OF HEAD IN FIGS. 21 AND 22. RAIN HAD PARTLY DIMMED BEARING SURFACE OF HEAD BEFORE TRAIN HAD PASSED OVER RAIL, AND THE SEVERAL WHEEL CONTACTS ONLY MADE BRIGHT BAND ABOUT 0.5 IN. WIDE. CRACK ON GAGE SIDE OF HEAD, SIMILAR TO THAT IN FIG. 18, WAS FOUND BY TRACKMEN AND RAIL REMOVED BEFORE IT BROKE.

The longitudinal fissure, and sometimes the vertical, open on the gage side of the head and can be discovered by careful inspection before rupture occurs in the track. This type develops in rails, or part of a rail, which cooled high on the hot beds and were then gagged upon the head to shorten its metal and lengthen that of the base below the neutral surface of the section. See Figs. 18, 19, 21, 22, 23.

The characteristic fractures are the same as the last two of the preceding type. Besides, the first may appear as a minute crack. The nucleus checked between the grains of the metal of the longitudinal fissure by the gag with its imprint is not visible to the trackmen, but will be disclosed by uncapping the head of the rail.

The illustrations of the coalescent type show typical forms of fractures that have developed in the track from different brands and sections of rails. The general features are the same though there will be variations

in minute details, owing to the various designs of the head blocks of the striking rams in the straightening presses. It also makes a difference whether the rail straighteners apply too light, too heavy, or the proper blow of the gag on the head of the high rails to straighten them. When the blow is too light, a heavier second blow must be given practically in the same place to put the required permanent set in the surface of either a high or low rail.

It is a frequent occurrence in the mill that too heavy a blow is given by the gag on the head, and then the rail is turned base up for a light blow to correct the effect of the excessive blow on the head. When too heavy a blow of the gag is applied to the base of a low rail, it must be turned head up and given a light blow to reduce a part of the set of the heavy blow. Rails are sometimes completely fractured in the straightening press by the application of the blow from the gag. An interior core of brittle metal is the initial point of fracture in nearly every instance, and is found in the general location of the nuclei of interior transverse fissures.

Occasionally rails are heard to snap under the straightening press, in which case the interior non-ductile core of brittle metal is checked and broken, while the surrounding envelope of more ductile metal is elongated, or shortened as required, without injury, unless the applied blow is sufficient to rupture the entire rail section. Such rails would only be noticed and rejected by either the railroad or mill inspector when near the straightening press as the sound would be the only thing that would attract his attention.

We have found two rail-heads that contained longitudinal fissures caused by the blow of the gag on the head, the precursor of the coalescent type, which had above it, in each rail-head, a small interior transverse fissure of the intergranular type, of which the nucleus was checked by a blow of the gag on the base.

I am aware, by conversations and correspondence, that it is difficult for many conscientious investigators who are seeking for definite causes of the interior transverse fissures to comprehend the first effect produced by the gag on the occasional rail, and they must find a cause themselves before they can accept it, even when familiar in general with the manufacture of rails. This meager knowledge is insufficient for the investigation of interior transverse fissures, when a minute detail in the metal of the rail-head can be the inducing factor of final rupture.

The statistics of the 882,237 tons that I have collected would make at least double that number of rails, or 1,764,747, and in these, 1113 interior transverse fissures developed in the two types. This makes one occasional rail in 1585, which shows how difficult it is to find the occasional rail at the mill to make an investigation upon for the cause of interior transverse fissures before the rails have been placed in the track and the

service shows those that are defective. The investigator must trace back from the effects which he sees in the rail fractures to find the causes of the interior transverse fissures, and this work should include mill conditions as well as service in the track. Many investigators have failed to find, from the effects, the causes that produced the interior transverse fissures in the occasional rail-head.

The interior transverse fissures in the occasional rail-head seem, at first, so mysterious to most investigators that many give up, or else guess at a cause before they find the correlations between the first three effects, which they see in the final ruptures. The three separate causes, as previously indicated, cannot be direct or combined as a continuous fracture of the rail at one place and time, but the first effect induces the two succeeding, therefore it requires a year or several years of service in the track to develop the final rupture. This involves two periods of mechanical work for the occasional defective rail as a girder: (1) The manufacture and finish under the straightening press and (2) laid in the track to carry the wheel loads. There are, therefore, two different places of service in two separate periods of time.

There are three distinct kinds of fractures and separations of the metal in the rail-head in the final rupture due to the development of the induced interior transverse fissures, which are the indelible records of fractures or separations at three definite dates. (1) The direct fracture of the nuclei of the intergranular type, or of the longitudinal imprint of of the gag of the coalescent type; (2) the growth by slow detail of the specular surfaces in the track; (3) the final rupture, in the track, of the remaining metal of the rail-head and section.

These may be termed the progressive steps, in the logical order of development, of induced interior transverse fissures in the rare abnormal rail-heads of rail service.

The three preceding paragraphs explain the confusion of the Government investigators, for the induced interior transverse fissures were not properly understood by them and they assumed, without adequate investigation, the theory that they were caused by the wheel loads in the first case they investigated. This was discussed before the Institute at the New York meeting last February. Since their first report they have called them "fatigue fractures of the metal" and imply that all rails are subject to them under the present wheel loads.

The railroad officials reject both of these statements for they have knowledge of facts from their service records of rails which prove: That they are not fatigue fractures of metal; that all rails of normal metal are not subject to them under the present wheel loads; that it is only the occasional rail of abnormal metal in which they occur.

The Government investigators do not explain the origin, time, and place of occurrence of the nuclei, nor their variations in size from 0.0625

to 0.375 in. in diameter of the type of interior transverse fissures classified by me as intergranular. All fractures of this type show nuclei of considerable area, and not points, which is the common type in fatigue fractures. These areas, constituting the nuclei ruptures as units and not by slow detail, show positively and conclusively that fracture was caused by the application of a rapidly applied blow of a pressure

FIG. 24.

beyond the elastic limit, or even ultimate strength, of this interior core of inferior metal.

I cite from a report by the Division of Safety of the Interstate Commerce Commission, as an exhibit of the application of the gag upon the head of the occasional rail, which accompanies the report of the accident on the Central of Georgia Railway, near Juniper, Ga., on Oct

30, 1917, and issued May 2, 1918. "Fig. 24 shows a group of four rail-heads in each of which progressive shearing fractures developed, the origins of which were at longitudinal streaks in the steel." These are longitudinal fissures with end shearing fractures, which follow the stresses in the metal from the applications of the gag upon the heads of the high rails, and enlarged by the pressures of the wheel loads rolling over them. The imprint of the gag is visible on each of the four specimens though more distinct in the two upper than in the two lower. The metal in the head of the rail, as already described, must be upset by the blow of the gag to shorten the head of the rail while the base must be lengthened by the same blow.

The summary of that report states: "Many features connected with their occurrence have been assented to by common concurrence of opinion." This does not apply to the nuclei, their origin and fracture, for the Government investigators have not explained their occurrence, though the growth or development of the interior transverse fissures around the checked nuclei occur as practically stated in their reports. "But notwithstanding this, remedial efforts have not been inaugurated and no adequate protection has been provided against the recurrence of this type of fracture."

This is not considered a correct or even a fair statement by railroad officials for the reason that a number of steel plants making basic open-hearth rails continued their former mill practice of reheating the blooms to roll the rails. I have given the figures collected from the service records to prove that rails rolled from reheated blooms rarely develop any interior transverse fissures, while they are abundant in rails that are rolled direct. One mill, as already stated, has introduced reheating furnaces for its output and since then there has been a radical reduction of failures of this type. The mills are now giving more attention to the hot-bed treatment and regularity of practice than was the case in the early days of the manufacture of basic open-hearth rails. Several of the plants have introduced straightening presses with 60-in. supports, which reduces the pressure upon the 105-lb. rail sections by 50,000 lb. to put a permanent set in the metal to straighten the rail, as compared to the pressure required in straightening this section on 42-in. supports. Furthermore, the above sentence from the Government report is uncalled for, as the railroad officials have taken the necessary steps in the past years to find out and study the actual causes of the induced interior transverse fissures in rail-heads.

Rails rolled from reheated blooms have shown greater deflections under the drop than direct-rolled rails. I made the first low-phosphorus and high-carbon Bessemer rails in 1890 where the blooms were reheated. A study was made of the permanent sets in the drop test of the 75- and 80-lb. section, and from the comparison of the moments of inertia of the

respective larger rail sections, I determined what the deflections should be for them, and also the chemical composition for the larger sections, which required an increase in carbon as the weight was augmented. This will explain the frequent question why it is necessary to increase the carbon for the larger sections of rails when rolled from the same size of ingots. It is to meet the loss in physical properties, for the reduction of the metal would be less in the larger section. The mills were without experience in rolling the large sections and were unable to suggest a chemical composition that would insure the proper physical properties in the larger sections. I also found that the deflections of the rails were slightly greater from reheated blooms in the larger heads of the 95- and 100-lb. section, in which the widths of the heads were the same, although the former was 0.0625 in. less in depth. The resistance to wear of these rails in subsequent years was slightly better for the reheated blooms than for the direct-rolled rails under the same wheel loads. The ductility in the bearing surface did not become exhausted as much as in the case of the rails rolled direct from the ingots. Many of the 95-lb. rails were in service until 1917, and were not worn out, though outclassed as girders. The 95-lb. and 100-lb. Bessemer rails were made of 0.06 per cent. phosphorus and 0.56 to 0.65 per cent. carbon. They were all rolled direct after 1892 to 1898. The comparisons of the relative wear were made with rails rolled from reheated blooms in 1891–92. The 100-lb. rails were rolled direct in December, 1892, but were not laid in the track until 1893. The roadmasters, as well as myself, observed the difference in the wear of the rails in the yearly inspections of the track.

Rails from reheated blooms are provided with higher temperatures for rolling, then the cold day, delays, first of the round, bunched, or spaced on the hot beds are factors of small significance, compared to their greater significance for the occasional rail to be affected by direct rolling.

Drop Tests for Brittle or Ductile Metal.—There were 329 rails rolled direct that developed induced interior transverse fissures in the track, which have been subjected to drop tests to ascertain the brittle or ductile metal in the rail-heads. The characteristic location of brittle metal per rail-head separates the rails into three classes: 96 rails were brittle for their entire length; 101 rails were brittle only in the vicinity of the fissure, while the balance of each rail withstood several impacts of the falling tup without developing fracture, showing more than 4 per cent. ductility in the rolled bearing surface; 132 rails were brittle in a part of their length and the balance was ductile. Several rail-heads have shown non-ductile cores of metal in the interior of the rail-head, which instead of running longitudinally the entire length as brittle metal, changed to the opposite side of the head. See Fig 25.

My previous paper before the Institute[7] illustrated interior transverse fissures in which the nuclei of the intergranular type were composed of well-developed polyhedral crystals. We have broken 132 rails from different melts and split them into several pieces and have found crystals in 22. Grain growth to polyhedral crystals is the result of too long germinative temperature of manufacture, and not to subsequent wheel loads in the track. We have also found delayed transformations in several rail-heads, which were confirmed by heating and cooling curves made on small pieces taken near the interior transverse fissures. Abnormal physical conditions of metal in the interior of the rail-heads were confirmed by tensile tests and others by magnetic tests.

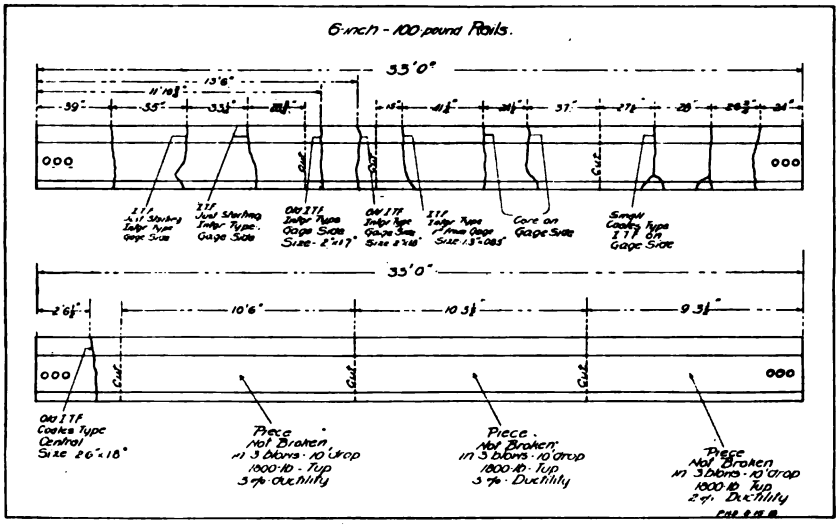


FIG. 25.—DROP TESTS FOR BRITTLE OR DUCTILE METAL IN RAIL-HEADS IN WHICH INTERIOR TRANSVERSE FISSURES HAD DEVELOPED. UPPER RAIL, BRITTLE ENTIRE LENGTH. LOWER RAIL, BRITTLE ONLY IN VICINITY OF FISSURE.

We have recently conducted extensive tests and surveys on rails by means of an improved type of magnetic rail tester without the necessity of breaking or destroying the material. The leakage curves of rails by this testing apparatus furnish a fund of information about the physical properties in rails and the serious disturbance of the metal in the heads and bases of rails by the gag of the press to straighten the rails in the finishing department of the mills.

Stremmatograph test No. 15, N. Y. C. & H. R. R. R., 5-in. 80-lb. rail, ties 15 in. centers, train No. 5, taken Sept. 30, 1897, is shown in Fig. 26. The 5-in. 80-lb. rail was straightened on supports 30 in. apart, which made the surface of the rails wavy and increased not only the train resistance but the unit fiber stresses in the base of the rail. Speed was 40 mi. per hr. and under rapid acceleration. The features of the short

[7] *Trans.* (1918) **58**, 627.

P. H. DUDLEY'S STEREOGRAPHOGRAPH TESTS, WEST ALBANY, N.Y., SEPT. 30, 1897.

No.15.- New York Central & Hudson River Railroad, 5-inch 80-lb. inside rail, ties 15 inch centers. Train No. 5. Locomotive No. 888, Class I, drawing five Wagner palace cars, all 6-wheel trucks. Speed 40 miles per hour, under rapid acceleration.
The extensions of the metal in the base of the rail, due to the several wheel loads of locomotive and cars can be traced, those of the locomotive being very distinct, while the compression of the metal between the wheels can also be seen. The rising of the rail in front of the front truck wheel, is plainly seen on the records.

Location of Stresses	Per Wheel	Per wheel base of Drivers and Trucks of Engine, Tender and Cars.	Per wheel base of Engine, Tender and Cars.	Per wheel base of Locomotive and Cars.	Per Pound of Wheel Base of Drivers and Trucks of Engine, Tender and Cars.	Wheel effects. Positive bending under wheels inch pounds.	Wheel effects. Negative bending moments between wheels inch pounds.
Locomotive — Engine — Front Truck — Compression in front of truck wheel	1,417	—	—	—	—	—	14,921
Tension under front truck wheel	16,070	—	—	—	—	169,217	—
Compression between wheels front truck	1,068	35,852	—	—	1.659	—	32,316
Tension under rear wheel front truck	13,576	—	—	—	—	132,457	—
Compression between rear truck wheel and front driver	5,433	—	99,746	—	—	—	57,209
Tension under front driver	31,415	—	—	—	—	330,800	—
Engine — Driving Wheels — Compression between front and rear drivers	6,126	63,894	—	—	1.589	—	22,387
Tension under rear driver	26,456	—	—	—	—	278,582	—
Compression between rear driver and front tender wheel	2,362	—	—	159,677	—	—	24,872
Tension under front tender wheel	12,755	—	—	—	—	134,310	—
Tender — Front Truck — Compression between wheels front truck	1,181	29,761	—	—	1.470	—	12,436
Tension under rear wheel front truck	11,465	—	—	—	—	121,765	—
Compression between front and rear trucks	2,362	—	58,731	—	—	—	24,872
Rear Truck — Tension under front wheel rear truck	12,591	—	—	—	—	132,595	—
Compression between wheels rear truck	1,889	29,170	—	—	1.440	—	19,891
Tension under rear tender wheel	12,755	—	—	—	—	134,310	—
Compression between rear tender wheel and front car wheel	709	—	—	—	—	—	7,466
First Car — Front Truck — Tension under front car wheel	14,408	—	—	—	—	151,716	—
Compression between front and middle wheels	1,181	—	—	—	—	—	12,436
Tension under middle wheel	14,177	47,254	47,254	—	—	149,231	—
Compression between middle and rear wheels	3,445	—	—	—	—	—	36,255
Tension under rear truck wheel	13,224	—	—	—	—	139,248	—
Compression in rear of wheel	472	—	—	88,588	—	—	4,970
Compression in center of space between trucks	0	—	—	—	—	—	0
Compression in front of first truck wheel	945	—	—	—	—	—	9,951
Rear Truck — Tension under front truck wheel	11,437	—	—	—	—	120,379	—
Compression between front and middle wheels	2,362	—	—	—	—	—	24,872
Tension under middle wheel	12,362	41,354	41,354	—	—	129,329	—
Compression between middle and rear wheels	1,181	—	—	—	—	—	12,436
Tension under rear wheel	12,040	—	—	—	—	126,844	—
Compression between trucks of first and second cars	2,362	—	—	—	—	—	24,872
Second Car — Front Truck — Tension under front wheel of truck of second car	9,664	—	—	—	—	101,972	—
Compression between front and middle wheels	1,181	—	—	—	—	—	12,436
Tension under middle wheel	14,177	43,983	43,983	—	—	149,231	—
Compression between middle and rear wheels	2,126	—	—	—	—	—	22,387
Tension under rear wheels	11,461	—	—	—	—	121,765	—
Compression in rear of wheel	2,126	—	—	—	—	—	22,387
Compression between trucks, center of space	0	—	—	88,912	—	—	0
Compression in front of wheel of rear truck	3,073	—	—	—	—	—	32,337
Rear Truck — Tension under front wheel, rear truck	12,040	—	—	—	—	126,844	—
Compression between front and middle wheels	1,181	—	—	—	—	—	12,457
Tension under middle wheel	11,227	44,909	44,909	—	—	118,280	—
Compression between middle and rear wheels	2,126	—	—	—	—	—	22,387
Tension under rear wheel second car	11,810	—	—	—	—	124,359	—
Compression between trucks of second and third cars	3,071	—	—	—	—	—	32,337
Third Car — Front Truck — Tension under front wheel of third car	14,172	—	—	—	—	149,231	—
Compression between front and middle wheels	1,653	—	—	—	—	—	17,406
Tension under middle wheel	15,569	50,427	50,427	—	—	164,152	—
Compression between middle and rear wheels	708	—	—	—	—	—	7,455
Tension under rear wheel	14,744	—	—	—	—	154,201	—
Compression back of rear wheel	1,893	—	—	—	—	—	19,901
Compression in center of space between trucks	472	—	—	87,628	—	—	4,970
Compression in front of wheel of rear truck	2,126	—	—	—	—	—	22,387
Rear Truck — Tension under front wheel of rear truck	9,445	—	—	—	—	99,487	—
Compression between front and middle wheels	1,653	—	—	—	—	—	17,406
Tension under middle wheel	9,446	37,201	37,201	—	—	99,487	—
Compression between middle and rear wheels	3,071	—	—	—	—	—	32,337
Tension under rear wheel third car	9,920	—	—	—	—	104,458	—
Compression between trucks of third and fourth cars	2,598	—	—	—	—	—	27,357
Fourth Car — Front Truck — Tension under front wheel fourth car	14,172	—	—	—	—	149,231	—
Compression between front and middle wheels	1,890	—	—	—	—	—	19,901
Tension under middle wheel	17,754	49,226	49,226	—	—	134,300	—
Compression between middle and rear wheels	2,834	—	—	—	—	—	29,737
Tension under rear wheel front truck	13,227	—	—	—	—	139,280	—
Compression back of rear wheel	2,834	—	—	—	—	—	29,737
Compression in center of space between trucks	472	—	—	92,096	—	—	4,970
Compression in front of rear truck wheel	2,362	—	—	—	—	—	24,872
Rear Truck — Tension under front wheel rear truck	10,395	—	—	—	—	109,438	—
Compression between front and middle wheels	2,834	—	—	—	—	—	29,842
Tension under middle wheel	11,574	42,870	42,870	—	—	121,874	—
Compression between middle and rear wheels	2,834	—	—	—	—	—	29,842
Tension under rear wheel fourth car	10,157	—	—	—	—	106,953	—
Compression between trucks of fourth and fifth cars	4,960	—	—	—	—	—	52,229
Fifth Car — Front Truck — Tension under front wheel fifth car	16,554	—	—	—	—	174,105	—
Compression between front and middle wheels	1,653	—	—	—	—	—	17,406
Tension under middle wheel	15,117	55,506	55,506	—	—	159,182	—
Compression between middle and rear wheels	1,653	—	—	—	—	—	17,406
Tension under rear wheel front truck	15,876	—	—	—	—	166,637	—
Compression in back of wheel	2,126	—	—	—	—	—	22,387
Compression in center of space	236	—	—	99,580	—	—	2,484
Compression in front of wheel	2,126	—	—	—	—	—	22,387
Rear Truck — Tension under front wheel rear truck	10,395	—	—	—	—	109,438	—
Compression between front and middle wheels	1,653	—	—	—	—	—	17,406
Tension under middle wheel	12,046	43,814	43,814	—	—	126,844	—
Compression between middle and rear wheels	1,653	—	—	—	—	—	17,406
Tension under rear wheel fifth car	14,408	—	—	—	—	151,716	—
Compression in rear of wheel	1,417	—	—	—	—	—	14,921

Instrument returned to zero.

FIG. 26a.

Reduction of sum of Positive and Negative Bending Moments to one pound of static load, for one rail, under locomotive.

Engine	Sum of positive bending moments	911,056.00	inch pounds		
—	— of P.B.M. per pound of static load	14.17	—	—	
—	— of negative bending moments	139,869.00	—	—	
—	— of N.B.M. per pound of static load.....	2.12	—	—	
Tender	— of positive bending moments	547,180.00	—	—	
—	— of P.B.M. per pound of static load	13.68	—	—	
—	— of negative bending moments	73,369.00	—	—	
—	— of N.B.M. per pound of static load	1.83	—	—	
Locomotive .	— of positive bending moments	1,458,236.00	—	—	
—	— of P.B.M. per pound of static load	14.58	—	—	
—	— of negative bending moments	212,657.00	—	—	
—	— of N.B.M. per pound of static load	2.13	—	—	
—	— of wheel effects	1,670,873.00	—	—	
First Car ..	— of positive bending moments	815,747.00	—	—	
— — ..	— of negative bending moments	117,089.00	—	—	
Second — ..	— of positive bending moments	781,451.00	—	—	
— — ..	— of negative bending moments	143,996.00	—	—	
Third — ..	— of positive bending moments	771,000.00	—	—	
— — ..	— of negative bending moments	151,709.00	—	—	
Fourth — ..	— of positive bending moments	761,970.00	—	—	
— — ..	— of negative bending moments	808,041.00	—	—	
Fifth — ..	— of positive bending moments	687,960.00	—	—	
— — ..	— of negative bending moments	157,918.00	—	—	

NOTE: The 5-inch 80-lb. rails were all straightened on supports 30 inches apart and were rough. The dynamic impact of the wheel or "wheel effects" were large. The whels in rear truck, third car, were new, while the others had been in service six to eighteen months.

FIG. 26b.

P. H. DUDLEY'S STEREOGRAPHIC TESTS NOVEMBER 3rd, 1904.

Engine 883, Westbound, 5 cars. Speed 25 miles per hour. Single track, 4-5/8", 65-lb. rail. Delvay, N.Y., - Syracuse to Auburn, Nov. 3rd,1904. Weight of Engine, 180,000 lbs.; weight of Locomotive, 200,000 lbs.; weight on track, 40,000 lbs.; on drivers, 80,000 lbs.; weight of tender, 80,000 lbs.

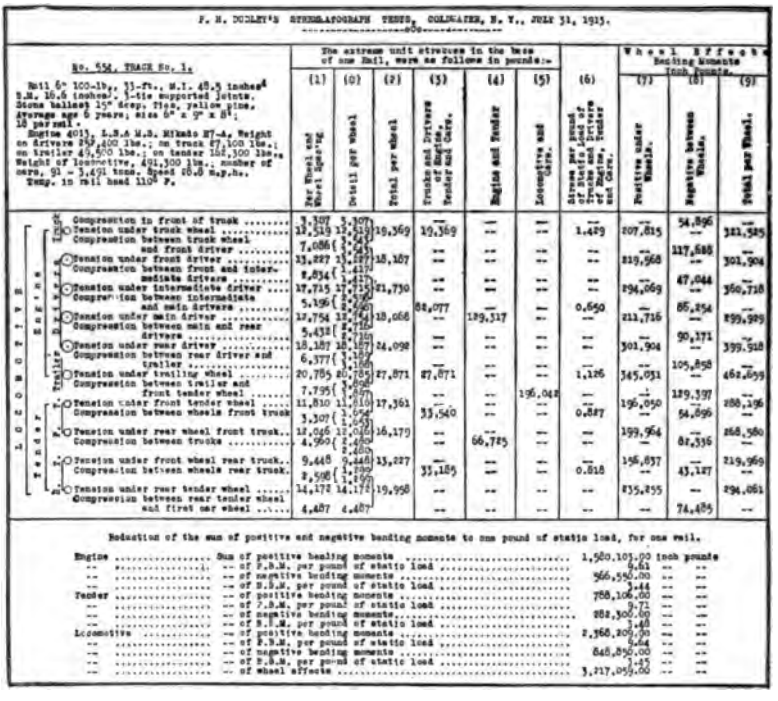
Location of Stresses.	Per Wheel	Per wheel base of			Per Pound of Wheel Base of	Wheel effects	
		Drivers and Trucks of Engine, Tender and Cars.	Engine, Tender and Cars.	Locomotive and Cars.	Drivers and Trucks of Engine, Tender and Cars.	Positive Bending Moments under Wheels, Inch-pounds.	Negative Bending Moments between Wheels, Inch-pounds.
LOCOMOTIVE — Engine — Truck — Compression in front of truck	1,563	--	--	--	--	--	29,513
Tension under front truck wheel	16,061	--	--	--	--	133,788	--
Compression between truck wheels	7,126	42,043	--	--	2.102	--	17,710
Tension under rear truck wheel	17,951	--	--	--	--	149,532	--
Compression between rear truck wheel and front driver	4,724	--	108,295	--	--	--	39,351
Engine — Drivers — Tension under front driver	34,656	--	--	--	--	291,183	--
Compression between drivers	7,854	66,252	--	--	1.656	--	23,607
Tension under rear driver	24,092	--	--	--	--	200,686	--
Compression between rear driver and front tender wheel.	4,015	--	--	175,020	--	--	33,445
Tender — Tension under front tender wheel	16,061	--	--	--	--	133,788	--
Compression between wheels front truck.	709	33,894	--	--	1.695	--	5,906
Tension under rear wheel front truck...	14,172	--	66,725	--	--	118,053	--
Compression between trucks	1,890	--	--	--	--	--	15,744
Tension under front wheel rear truck...	14,644	--	--	--	--	121,985	--
Compression between wheels rear truck..	236	32,832	--	--	1.642	--	1,966
Tension under rear tender wheel	15,825	--	--	--	--	131,822	--
Compression back of rear tender wheel..	2,362	--	--	--	--	--	19,675
First Car — Tension under first car wheel	20,549	--	--	--	--	171,175	--
Compression between first and second car wheels	236	37,319	--	--	--	--	1,966
Tension under second car wheel	15,353	--	--	--	--	127,890	--
Compression,- center of wheel spacing..	0	--	77,945	77,945	--	--	0
Compression,- front of rear truck.....	2,598	--	--	--	--	--	21,641
Tension under third car wheel	22,439	40,626	--	--	--	186,917	--
Compression between third and fourth car wheels	709	--	--	--	--	--	5,906
Tension under fourth car wheel	14,680	--	--	--	--	123,950	--

Reduction of sum of Positive and Negative Bending Moments to one pound of static load, for one rail.

Engine	Sum of positive bending moments	775,189.00	inch-pounds	
—	— of P.B.M. per pound of static load	12.92	—	—
—	— of negative bending moments	126,904.00	—	—
—	— of N.B.M. per pound of static load	2.12	—	—
Tender	— of positive bending moments	505,648.00	—	—
—	— of P.B.M. per pound of static load	12.64	—	—
—	— of negative Bending moments	50,176.00	—	—
—	— of N.B.M. per pound of static load	1.25	—	—
Locomotive	— of positive bending moments	1,280,837.00	—	—
—	— of P.B.M. per pound of static load	12.80	—	—
—	— of negative bending moments	177,080.00	—	—
—	— of N.B.M. per pound of static load	1.77	—	—
—	— of wheel effects	1,457,917.00	—	—

FIG. 27.

span supports in the straightening press had been noticed in the track indicator diagram, which were repeated year after year on the same rails. It was found necessary to improve the surface of the rail by lengthening the supports to 40 or 42 in. which made a smoother surfaced rail. The opposite track was laid with the 5⅛-in. 80-lb. section and it was noticed that the same locomotive which would produce high stresses on the 5-in. 80-lb. rail would have much lower unit stresses at 50 mi. per hr. on the smooth 5⅛-in. 80-lb. rail. The wheels on all of the trucks under the palace cars except the third car, rear truck, had been in service from 14 to 18 mo. and were slightly rough upon the treads. The wheels



F. H. DUDLEY'S STRESS-AUTOGRAPH TESTS, COLDWATER, N. Y., JULY 31, 1913.

No. 454, TRACK No. 1.

Rail 6" 100-lb., 33-ft., M.I. 48.5 inches⁴ S.M. 10.6 inches³, 3-tie supported joints. Stone ballast 15" deep. Fine, yellow slime. Average age 6 years; size 6" x 9" x 8'; 18 per rail.

Engine 4013, L.S.& M.S. Wheels 67-4. Weight on drivers 219,400 lbs.; on truck 67,100 lbs.; on trailer 49,800 lbs.; on tender 102,300 lbs. Weight of locomotive, 491,300 lbs.; number of cars, 9; = 1,401 tons. Speed 50.0 M.p.h. Temp. in rail head 110° F.

The extreme unit stresses in the base of one Rail, were as follows in pounds:-

	(1) Per Wheel and Wheel Spacing	(2) Details per Wheel	(3) Total per Wheel	(4) Trucks and Drivers of Engine, Tender and Cars.	(5) Engine and Tender	(6) Locomotive and Cars.	(6) Stress per pound of Static load on Engine, Tender and Cars.	Wheel Effects Bending Moments Inch Pounds. (7) Positive under Wheels.	(8) Negative between Wheels.	(9) Total per Wheel.
Engine: Compression in front of truck	3,307	3,307								
Tension under truck wheel	12,519	12,519	19,369	19,369	—	—	1.429	207,815	54,896	321,525
Compression between truck wheel and front driver	7,086	7,543								
Tension under front driver	13,227	13,227	18,187	—	—	—	—	219,568	117,688	301,904
Compression between front and intermediate drivers	2,834	2,417								
Tension under intermediate driver	17,715	17,715	21,730	—	—	—	—	294,069	47,044	360,738
Compression between intermediate and main drivers	5,196	2,566		84,077			0.650		85,254	
Tension under main driver	17,754	18,066	18,066	—	189,317	—	—	211,716	—	299,989
Compression between main and rear drivers	5,432	2,735							90,171	
Tension under rear driver	18,187	18,187	24,098	—	—	—	—	301,904	—	399,918
Compression between rear driver and trailer	6,377	3,166							105,858	
Tension under trailing wheel	20,785	20,785	27,871	27,871	—	—	1.126	345,051	—	462,699
Compression between trailer and front tender wheel	7,795	3,958		—	—	196,048	—	—	129,397	
Tension under front tender wheel	11,810	11,810	17,361	33,540	—	—	0.827	196,050	—	288,196
Compression between wheels front truck	3,307	1,552							54,896	
Tension under rear wheel front truck	12,046	12,046	16,179	—	66,725	—	—	199,964	—	268,580
Compression between trucks	4,969	2,460							82,536	
Tension under front wheel rear truck	9,448	9,448	13,227	33,185	—	—	0.818	156,837	—	219,969
Compression between wheels rear truck	2,598	1,779							43,147	
Tension under rear tender wheel	14,172	14,172	19,958	—	—	—	—	235,255	—	294,081
Compression between rear tender wheel and first car wheel	4,487	4,487		—	—	—	—	—	74,485	—

Reduction of the sum of positive and negative bending moments to one pound of static load, for one rail.

Engine	Sum of positive bending moments	1,580,103.00	inch pounds	
"	" of P.B.M. per pound of static load	8.61	"	"
"	" of negative bending moments	566,550.00	"	"
"	" of N.B.M. per pound of static load	3.44	"	"
Tender	" of positive bending moments	788,106.00	"	"
"	" of P.B.M. per pound of static load	9.71	"	"
"	" of negative bending moments	282,300.00	"	"
"	" of N.B.M. per pound of static load	3.48	"	"
Locomotive	" of positive bending moments	2,368,209.00	"	"
"	" of P.B.M. per pound of static load	8.64	"	"
"	" of negative bending moments	848,850.00	"	"
"	" of N.B.M. per pound of static load	3.45	"	"
"	" of wheel effects	3,217,059.00	"	"

FIG. 28.

in the rear truck of the third car had only been in service about 3 mo. and showed a reduction in the unit stresses in the rail.

The tests of Nov. 3, 1904, shown in Fig. 27, were upon 4⅝-in. 65-lb. rail at Solvay, N. Y. The unit stresses in these light rails were higher than in the 5⅛-in. 80-lb. rail for the same locomotive and speed, as would be expected.

The stremmatograph tests at Coldwater, N. Y., July, 1913, shown in Figs. 28 and 29, indicate by their serial numbers that several hundred had been taken from 1897 to the latter date, to see that the locomotives and cars were not exceeding the stresses suitable for the rails. When

this was found to be the case the speed was limited, which practically meant the replacement of the light rails by those of stiffer sections. There are several tabulations of tests on the tracks at Coldwater. Fig. 28 has an additional column to explain in detail how the figures are obtained for the division of the stresses per wheel. Column (0) is only introduced

P. H. DUDLEY'S STREMMATOGRAPH TESTS, JULY 31ST, 1913.

No.555.- Track No.1, Cold Water, N.Y.. Rail 6 inch, 100-lb.. Moment of inertia, 48.5 4th power inches. Length 33 feet, joints, 3-tie supported, 36 inches long. Ballast, stone 15" deep. Ties, yellow pine. Average age, 6 years. Size 6" X 9". Number per rail, 18. Engine No.3740, N.Y.C.& H.R.R.R.. Mikado type H-7-a. Weight on drivers, 215,000 lbs.; on truck, 24,000 lbs.; on trailer, 45,000 lbs.. weight of tender, 154,500 lbs.; weight of locomotive, 438,500 lbs.. Number of cars, 91. Speed 51.5 miles per hour. Temperature in mill head, 110 deg. F..(New Rail)

Location of Stresses.	Per Wheel	Per Wheel Base of: Drivers and Trucks of Engine and Tender.	Per Wheel Base of: Engine and Tender	Per Wheel Base of: Locomotive and Cars.	Per Pound of Wheel Base of: Drivers and Trucks of Engine and Tender.	Wheel Effects: Positive bending moments under wheels, inch-pounds.	Wheel Effects: Negative bending moments between wheels, inch-pounds.
LOCOMOTIVE: Compression in front of truck	2,598	--	--	--	--	--	43,127
Engine: Front truck: Tension under front truck wheel	6,577	11,537	--	--	0.945	105,858	--
Compression between truck and front driver	4,774	--	--	--	--	--	78,418
Tension under front driver	26,090	--	--	--	--	443,054	--
Compression between front and intermediate drivers	5,196	--	--	--	--	--	86,254
Tension under intermediate driver	23,147	--	--	--	--	384,240	--
Compression between intermediate and main drivers	2,834	101,800	--	--	0.947	--	47,044
Tension under main driver	19,604	--	135,575	--	--	325,426	--
Compression between main and rear drivers	2,126	--	--	--	--	--	35,292
Tension under rear drivers	17,951	--	--	--	--	297,987	--
Compression between rear driver and trailer	3,779	--	--	--	--	--	62,731
Engine: Trailer truck: Tension under trailing wheel	18,187	22,458	--	--	1.00	301,904	--
Compression between trailer and front tender wheel	4,774	--	--	193,206	--	--	78,418
Tender: Tension under front tender wheel	10,629	--	--	--	--	176,441	--
Compression between wheels, front truck.	1,953	31,178	--	--	0.807	--	27,440
Tension under rear wheel, front truck..	14,880	--	--	--	--	247,008	--
Compression between trucks	3,307	--	57,631	--	--	--	54,896
Tension under front wheel, rear truck..	13,699	--	--	--	--	227,405	--
Compression between wheels, rear truck..	2,126	26,455	--	--	0.685	--	35,292
Tension under rear tender wheel	7,558	--	--	--	--	125,463	--
Compression back of rear tender wheel..	2,834	--	--	--	--	--	47,044

Reduction of sum of positive and negative bending moments to one pound of static load, for one rail.

Engine	Sum of positive bending moments	1,858,469.00	inch-pounds	
--	-- of P.B.M. per pound of static load	13.09	--	--
--	-- of negative bending moments	398,075.00	--	--
--	-- of N.B.M. per pound of static load	2.76	--	--
Tender	-- of positive bending moments	776,315.00	--	--
--	-- of P.B.M. per pound of static load	10.05	--	--
--	-- of negative bending moments	180,359.00	--	--
--	-- of N.B.M. per pound of static load	2.34	--	--
Locomotive	-- of positive bending moments	2,634,784.00	--	--
--	-- of P.B.M. per pound of static load	12.02	--	--
--	-- of negative bending moments	572,434.00	--	--
--	-- of N.B.M. per pound of static load	2.61	--	--
--	-- of wheel effect	3,207,218.00	--	--

FIG. 29.

in this tabulation but the key can be used to understand any of the other tests.

These stremmatograph tests show the care used to keep the unit stresses in the rails to proper limits as the wheel loads and total loads of the locomotives and cars have increased. The important principle has been used to subdivide the total load of the locomotives and increase the number of drivers for the tractive effort desired and at the same time keep the unit stresses within the desired figure for safety in operation.

The M. C. B. Contours for Wheel Treads.—The M. C. B. contours for cast-iron wheels and for solid steel tires for locomotives and tenders have too narrow wheel treads to distribute properly their loads to the metal of the bearing surface of the rail-heads. The New York Central Railroad, and some of the subsidiary lines, returned to a contour of 1 in 38 across the tread in 1912 for locomotive and passenger-car wheels, and in 1916 changed the contours of the cast-iron wheels to an inclination of 1 in 38 entirely across the tread to the sand rim. The M. C. B. contours for cast-iron wheels for freight cars deform the rail-heads down on the gage side and build them up square on the outside corner.

The areas of contact taken under a Pacific type locomotive on the deformed rail-heads of the 100-lb. section on Sept. 14, 1917, developed an average intensity of pressure for the 20 wheels of 82,150 lb. New 105-lb. rails were laid in this track May, 1918, and the experiment repeated Jan. 21, 1919, with a similar Pacific type locomotive, having N. Y. C. standard wheel contours and nearly new. The average intensity for the 20 wheels was only 57,425 lb. This was a reduction of over 43 per cent. in the average intensity of pressure, and therefore a better distribution of the loads to the metal in the bearing surface of the rail-head, as proved by the figures just stated. It will take some years to reëstablish the efficient correlations between the wheel treads and rail-heads that existed before the M. C. B. contours of 1909 were adopted. The correction has started already and will become general before many years.

Mr. Comstock, in connecting my name and work with the theory of fatigue fractures of metal in rail-heads, has done so without my approval. I do not see, in my constructive work of so many years and examination of so many rails after 25 years of service, the slightest evidence of fatigue fracture of the nuclei of interior transverse fissures under the present wheel loads. This is a case where I must see positive evidence before I can believe that these have occurred, and I am on the lookout for the public and the railroad side of the question all the time.

Conclusions.—The service records show that reheated blooms coupled with proper chemical composition and mill practice will practically eliminate interior transverse fissures in rail-heads of the occasional rail.

The service records show that investigations of the years past indicated the proper design and composition for the manufacture of rail sections in Bessemer and open-hearth steel, that have permitted on the same roadbed the driving wheels and their loads to be trebled and the car wheels and their loads to be quadrupled with large factors of safety in operation.

The service records show, by the general practice of the railroads, that the tensile stresses in the spacing of the wheel loads are so far below the elastic limits of the steel that the alternate stresses do not approximate those necessary for fatigue fractures of metal in the interior of the

rail-heads. This is shown in the stremmatograph records as a part of this paper before the Institute.

The contours of the wheel treads should return to those in use when the designs of the rail sections were made which enabled the increase in wheel loads. This is required to distribute the wheel contact pressure and loads to a sufficient width of metal in the bearing surface of the rail-head to hold the average intensity of pressure, practically constant in rail-heads of 3 in. width.

The law of the service records under the daily observation of the railroad officials is more reliable proof of the safety in transportation than an opinion that does not take cognizance of the facts.

THE CHAIRMAN (J. E. JOHNSON, JR., New York, N. Y.)—I should like to ask why it is, if these phosphorus streaks seem to account for the transverse fissure, that the open-hearth rail, which has half of the phosphorus that the old Bessemer had, has shown ten times as many transverse fissures per ton of rail produced?

G. F. COMSTOCK.—I think the possible explanation of that is that in the Bessemer steel we have an envelope of ferrite surrounding each pearlite grain, whereas in the open-hearth rail that ferrite is absent. You will find that point considered on page 716 in my paper.

G. M. DAVIDSON,* Chicago, Ill.—One interesting phase of the subject that has not been touched upon is, how are we going to find out about these rails after they are in the track? I know of a railroad that has been looking for transverse-fissure rails for several years, but found none until within the last six months, during which time it found about 30. All of these rails were rolled in the winter of 1912 and laid on the track in the spring of 1913, so it begins to look as though six years was about the time required to develop transverse fissures on this railroad. The interesting fact is that we do not know how to pick those rails out of a track. I have brought here a couple of samples that illustrate this. Here are two cross-sections cut from a rail that broke in the track on account of a transverse fissure. When a piece of this rail was picked up and carried on a truck it fell off and broke into three pieces. These two transverse fissures are from this same rail, about 6 ft. apart. One of them is rusty, and the other is bright and has a film of good metal all around it. Neither of these fissures could be seen until the rail broke. Now, how can a railroad determine when rails have transverse fissures and should be taken out of the track? It ought not to be necessary to wreck a train to find them. If anybody has any suggestion as to a method for detecting transverse-fissure rails in the track, I am sure the railroads would be very glad to know about it.

* Chemist and Engineer of Tests, C. & N. W. Ry. Co.

CHAIRMAN JOHNSON.—That is very much to the point. I think a good many people have spent time on this question of finding transverse fissures before they wreck the train. There has been some talk of the work done by the Bureau of Standards to prove discontinuity in the metal by means of magnetic tests but I think that has only advanced as far as telling something about the rail when it is run through a magnetic machine, which involves removing it from the track to test it, which, of course, is practically out of the question. In conjunction with the Robert Hunt Co. a few years ago, I did a little work to see if we could detect a change in the rail at one of the transverse fissures, by a change in the electrical resistance, but this had a negative result.

C. B. BRONSON,* New York, N. Y.—Mr. Comstock's paper is valuable and important because a scientific explanation is made for a long recognized practical fact. The superiority of rails rolled from reheated blooms and their freedom from interior transverse fissures was never more clearly comprehended and demonstrated than by our recent compilation of failures in direct rolled rails and those from reheated blooms, on eight heavy-traffic roads. This analysis points to the fact that 90 per cent. of the failures due to interior transverse fissures may be eliminated by reheating the blooms—a matter of vital importance to all railroads of this country.

The most prominent idea in the paper which stands out clearly may be summed up in one word, uniformity. The emphasis is upon phosphorus distribution which is considered to be the deciding factor for long service, or the early failure of the rail. This seems like a fine distinction to make when we consider that 0.03 per cent. is a liberal value for phoshorus in the average open-hearth rail analysis. The question then is: Does this limited amount of phosphorus exert such a large influence on the resultant product? Furthermore, considering the nature of this phosphorus distribution, in what way do these apparent coarse laminations affect the integrity of the steel? Are these laminations in the nature of cleavage or shear planes, or fibers of low strength, rigidity, and elasticity? Until these matters are decided, judgment of the conclusion presented should be withheld.

Uniformity pertains not only to this question concerning phosphorus streaks and distribution, but stands out in a very practical way. Close observation of numerous rollings of rails shows very markedly that mill practice, from the soaking pits to the finishing department, is decidedly more uniform where reheating the blooms is resorted to than where the "ingot to rail" method is followed. These advantages are readily noticeable:

1. Poor practice in blooming some ingots can be partly corrected by reheating.

*Of the New York Central R. R.

2. Thorough soaking in the reheater tends to normalize the badly crushed ingot structures, the resultant effect of the enormous pressure imparted in reducing the massive ingot structure to an 8 in. by 8 in. bloom.

3. The temperature of the bloom is increased about 300° F. as a result of 1 hr. soaking.

4. All blooms leave the reheater at nearly the same temperature, as generally they are reheated in one continuous furnace; in contrast to the soaking of ingots, which is done in many pits, with variable temperature regulation and control.

5. Higher finishing temperature for the rail flanges, and a more uniform temperature throughout the rail section.

6. A uniform and nearly constant finishing temperature for bloom after bloom.

7. The effect of the latter results in more uniform hot-bed practice and anneling conditions of the rails.

8. The destructive work of straightening is lessened as the final sweep of the cold rails is under closer control.

These practical advantages must be thoroughly and carefully considered as well as the question of the desirability of uniform phosphorus distribution. Bessemer steel, containing about four times as much phosphorus as open-hearth steel, rarely develops interior transverse fissures. Only four such instances are known on the New York Central Railroad in over 1,000,000 tons of rails of 0.10 per cent. phosphorus and 0.55 per cent. carbon, although some failures have developed in the early Bessemer rails rolled from horizontally reheated ingots. Some roads report a number of Bessemer failures of the high phosphorus composition, but a careful examination of the alleged fissures generally discloses this to be an error in describing them on the "failed-rail reports."

The absence of fissures in Bessemer steel is puzzling and contradictory to many, especially when viewed from the standpoint of the fatigue-failure theory. It has been constantly stated by advocates of this viewpoint that it makes no difference what is the rail composition, physical characteristics, or structures, fatigue can ultimately occur. We do not have the so-called fatigue fractures in Bessemer steel, though the hardness and wearing qualities of this class of rail is nearly equivalent to that of our present standard open-hearth steel. This fact alone casts considerable doubt on the rudiments of the fatigue fracture theory.

The second important feature of Mr. Comstock's valuable paper is an effort to reconcile the two main viewpoints of the cause of this type of fracture, and therefore calls for very careful consideration. The fundamental and basic principles of the fatigue theory were tersely mentioned by J. E. Howard, its strongest advocate, as early as 1908, in a paper discussing "Some Causes of Rail Failures," before interior

transverse fissures were common and generally known. This paper was presented about the time that the manufacture of basic open-hearth steel for rails was commenced on a quantity production basis. The catastrophe at Manchester in 1911, investigated by the Engineer-Physicist of the Interstate Commerce Commission, brought forth a report and conclusion which was practically a reiteration and extension of this previously expressed fatigue failure idea, but worded in stronger terms. Many railroad engineers were led to believe that thousands of rails were susceptible to failure of this type, if the conclusions presented were correct, and might lead to failures at such an alarming rate that service would be impaired and demoralized and safety would become difficult to maintain. These frank conclusions were introduced in that report without corroborative proof or evidence, merely being an expression of opinion.

Since that time, report after report has been issued by the Government reiterating the sense of the previously expressed statements and a long research conducted into the physics of rail steel from the standpoint of inherent and developed stresses, and an involved study of the mechanics of ultimate failure. Stresses have been shown to exist, generally of moderate or even meager magnitude, except those at the extreme topmost running surface or sides of the rail-head, and the occasional high stresses in the flanges, which have but little bearing on the subject. Stresses of a few thousand pounds, about 5000 lb. per sq. in. to be exact, have been found in the interior of the rail-head, and this together with the isolated compressive stresses existing only at the running surface and the tensile strains of service from wheel loading are held primarily responsible. The fact is lost sight of entirely that the tensile stresses of service in the rail-head do not occur under the wheels but between them. A study of thousands of measurements of service stresses under moving loads shows beyond a doubt that these tensile stresses developed in front or back of the passing wheels—which gives the rail its reverse bending—average about 7500 lb. per sq. in. for the topmost fibers of the rail-head, and rarely reach a value as high as 15,000 lb. per sq. in. The advocates of the fatigue-failure theory contend that this limited tensile stress from service coupled with the few thousand pounds of induced residual tensile stress in the interior of the rail-head—the effect of cold rolling the rail-head surface—is sufficient to fracture and fatigue the nucleus of the interior transverse fissure—an area of $\frac{1}{8}$ to $\frac{3}{8}$ in. in diameter.

Furthermore, effort has been made in successive Government reports to disprove the accumulated evidence gathered by the railroads, and confuse the issue by indicating great variance in the knowledge and opinions published or privately expressed by engineers and "surface" investigators. Not once, however, has a practical constructive solution been offered or tendered to aid the railroads in combating this intricate prob-

lem, even though the Government investigators have had every facility offered them for investigation. The nature of the work has been destructive, not constructive, in that a virtual recommendation has been made for a reduction in present wheel weights and loadings.

But in spite of this opposition and constant criticism, the railroads have gathered evidence, searched for causes, called for and insisted upon the recommended remedies, and their efforts have been amply rewarded by reduction of failures in the product of mills where poor practices have been eliminated and improved practices initiated. These are not mere expressions of opinion, but are based on minute examination, study, and research for over six years on failures of this type on the largest railroad system in the country. Statistics have been compiled, probably as complete as any that exist, analyzed and studied from every conceivable angle, operating conditions gone into thoroughly, evidence for and against every possible argument weighed and balanced, and out of this the true facts of the case have been evolved.

We have been criticised for our methods in tackling this difficult problem, but it is here stated with emphasis that no stone has been left unturned to get at the root of the trouble. Special testing methods have been resorted to, displacing old standards, thermal analyses made, special microscopic examinations instituted as well as other methods of physical testing, strain measurements, and even several series of magnetic surveys on full length rails and short isolated pieces, work of a very exact nature; proof enough that we have not neglected the duty that has developed upon us to formulate a solution. The marked reduction in the number of failures, as slight mill changes have been made and the causes of interior transverse fissures became understood, enunciated, and accepted, is testimony enough to the ability and foresight with which this problem has been encountered.

The rates of failures in the earliest open-hearth rollings, which were made from compositions high in hardening elements and under conditions that would not be tolerated at the present time, are all out of proportion compared to rollings of more recent date, the failures being so high as to be responsible for the greatest number of failures. As an instance, 72 per cent. of all interior transverse fissures are concentrated into early rollings representing a few thousand tons of steel, and widely scattered over hundreds of miles of track out of the 400,000 tons that have been rolled. The abnormally high rate of failure for some mills in the early output has been practically eliminated in their subsequent product after a few practical recommended changes had been instituted. This is evidence enough to disprove that the individual characteristics of the rail are of no importance in discussing this type of failure.

The life history of an interior transverse fissure involves three distinct phases before ultimate rupture of the rail section occurs. They

may be summarized as follows: 1. Interior brittleness or lack of ductility induced in the steel, or primarily from the hardening elements of the chemical constituents; (2) the nucleus, its nature, appearance, and method of fracture; (3) growth of the smooth, silvery surfaces that surround the nucleus.

It is generally and widely recognized by mill people, metallurgical engineers, investigators for the railroad interests, and others, that interior brittleness, whether confined to isolated spots in the rail length, to the entire length of the rail, or to several rails, is the fundamental and prime cause for interior transverse fissures. This brittleness furnishes the potential factor for checking a relatively large area of metal—the nucleus—and the subsequent development to complete failure. Brittleness may be encountered from several causes, chief of which are acceptance of material rolled from a composition with high percentages of hardening elements, principally carbon, high rolling speeds and low finishing temperatures, or rapid and improper cooling conditions for the rails on the hot beds. It must be remembered that the class of rail steel we are now obtaining is a modified form of self-hardening steel as the eutectoid composition is approached, and is sensitive to air hardening and induced brittleness under rapid rates of heat conduction. The vital importance of proper cooling conditions and protection of the rails upon the hot beds is clearly recognized. Vigilance and care in protecting the rails on the hot beds from exposure to gusts of air, especially in the winter time, and uniformity of cooling should be insisted upon. The difference between a good and a bad rail often depends on slight practical details, as has been shown by actual tests made.

The mechanics of the third phase, that is, the formation of the lustrous surfaces that surround the checked nucleus, are well understood and now universally recognized as being a detail growth and enlargement, resulting from the reversal of bending to which the rail is subjected when the thousands of wheels pass over it. The method of checking the nucleus is generally overlooked by most investigators, although it constitutes the heart and vital part of the interior transverse fissure. The reason for checking the nucleus furnishes the solution and the necessary link in the development of the interior transverse fissure and the subsequent failure of the rail. The advocates of the fatigue-fracture theory overlook, apparently, the importance of the nucleus—the whole fracture being summed up under the general term *fatigue*. Fatigue fractures commence from the weakening of individual crystals or at most isolated or small groups of crystals, practically starting from one point. No evidence whatever has been seen in the nucleus of fatigue of individual crystals, but the evidence does show that rupture of this interior core of $\frac{1}{8}$ to $\frac{3}{8}$ in. in diameter as a single unit was caused by some sharp sudden impact of high intensity; in other words, an overload. This severe loading, which

is necessary for fracture of the brittle nucleus, occurs when the rail is straightened. Measured strains on rails undergoing gagging indicate stresses of 100,000 lb. per sq. in., a value beyond the ultimate strength of the interior brittle metal of an occasional rail, which must be transmitted to the section, as a sharp, sudden impact or blow to affect the permanent set. That this mechanical work upon the rail is excessively severe, and seriously disturbs the physical properties of the part of the rail overstrained, has been readily demonstrated by magnetic surveys on a number of full length gagged rails. Violent oscillations are transmitted to the indicator point of the recording mechanism, demonstrating that the magnitude of the disturbance to the magnetic properties of the rail is great, and therefore the physical disturbance is likewise, for these two properties are directly interrelated.

The argument has been repeatedly advanced that the only effect of gagging on the rail is a disturbance of the surface mill scale. That this statement is incorrect has been often times proved. In several instances rails have actually been broken in two when undergoing the process of gagging; the fracture radiating from a non-ductile center in the general location of the nuclei of the interior transverse fissures. In other cases, rails rupture internally without complete fracture due to the toughness and resisting power of the surrounding more ductile metal.

It is a curious coincidence that the advocates of the fatigue-failure view have found it necessary, in order to prove their point, to resort to the destructive mechanical overloading that occurs in gagging a rail, and which is the destructive agency that checks the nuclei of interior transverse fissures. Fatigue failures are guaranteed to be developed at will in any rail, whether good or bad, we have been told. The history of manufacture of these artificial failures, their physical characteristics, or other properties have never been furnished with the reported breakages. We have thousands of rails constantly in the most practical test that has, or can be devised—service—and have found by watching the records of these thousands of rails, as well as removing hundreds of them for test purposes, that interior transverse fissures are not and cannot be developed in this vital practical service test, as these artificially developed failures would lead us to believe.

The above-mentioned cases are certainly convincing proof of the origin, and checking of the nucleus of interior transverse fissures in railheads. Provision must therefore be made to reduce, as much as possible, injurious gagging effects, and it is for this reason that lengthening the span between straightening press supports is advocated. Criticism has been made of the fairness and value of the samples submitted to Mr. Comstock but it is here stated with positiveness that the history of all rails, except Nos. 28, 29, and 30, furnished by the New York Central, is fully known from rolling conditions to the time of withdrawal from serv-

ice. The rails have all been under very heavy service, either main line or freight and have carried in general from 100,000,000 to 200,000,000 tons for the good service rails, and lesser amounts for the failed sections.

In conclusion, a restatement is made of a few practical points which must be definitely and carefully considered in the manufacture of all rail rollings: (1) Avoid the use of extremely high carbon. (2) Reheat all blooms before rolling them into rails. (3) Finish at a uniform and constant temperature. (4) Protect the annealing rails upon the hot beds. (5) Carefully camber all rails so that when cold they will be as nearly straight as it is possible to obtain them. (6) Reduce the intensity and destructive mechanical work of straightening, which is the deciding factor in the fracture of the nucleus of the interior transverse fissure. The observance of these few practical points has helped to make possible the carrying of millions of wheel loads yearly over hundreds of thousands of rails without fracture, or anticipation of fracture from failures of this type.

M. H. WICKHORST,* Chicago, Ill.—Mr. Davidson showed samples of a type of interior break that occurs in rails, called "transverse fissure;" illustrations of simple transverse fissures are given in Figs. 30 and 31. This type of fissure consists of a granular nucleus surrounded by a bright polished part, which has undoubtedly grown

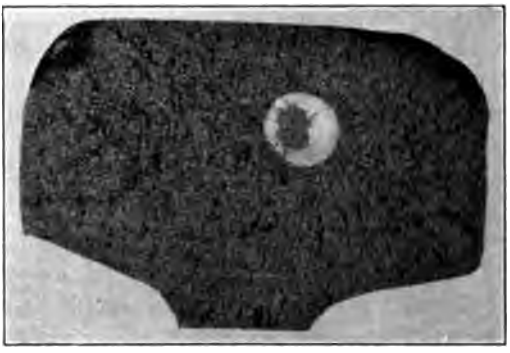


FIG. 30.—TRANSVERSE FISSURE SHOWING SMALL DEVELOPMENT AROUND NUCLEUS.

around the nucleus as a result of constant alternations of stress in service. In size, it may vary from a slight band on one side of the nucleus to an area almost the full section of the head. If the crack has reached the surface at the side of the head, the weathering turns it dark in color; the trackman then calls the fracture a "black heart." The examination of these fissures in their numerous stages, beginning with the plain undeveloped nucleus or "gray spot," suggests that the nucleus is an original crack in the rail from which the fissure developed in service. The nucleus seems also to be the same thing as the flake that is attracting so much attention in other classes of material.

* Engineer of Tests, American Railway Engineering Assn.

When a longitudinal section of the head of a transverse-fissure rail is prepared as indicated in Fig. 32, and the slab is deeply etched in hot commercial hydrochloric acid for 30 to 45 min., the surface representing

FIG. 31.—TRANSVERSE FISSURE SHOWING LARGER DEVELOPMENT AROUND NUCLEUS.

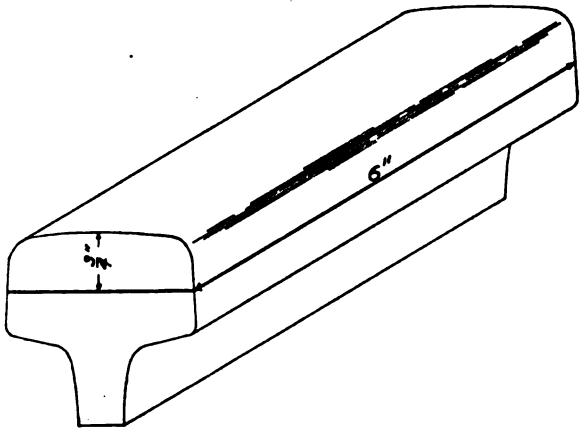


FIG. 32.—SLAB OF RAIL-HEAD FOR ETCHING.

FIG. 33.—ETCHED LONGITUDINAL HORIZONTAL SECTION THROUGH CENTER OF HEAD OF RAIL, SHOWING NUMEROUS SMALL INTERIOR CRACKS.

a horizontal longitudinal plane through the middle of the head has the appearance shown in Fig. 33. Here it is seen that the interior of the head contains numerous small cracks or lesions, an occasional one of which acts as a nucleus or center of growth for a fissure. It has been a "long, long trail" to develop our information concerning fissures to its present state, and the next step is to determine at what stage in the manufacture the small interior cracks, otherwise known as nuclei or flakes, are produced.

G. F. COMSTOCK (author's reply to discussion*).—The discussion of this paper has been of such volume and interest as to constitute in itself a sufficient excuse for the paper, even though the paper itself may have been as much in error as some of its critics have maintained.

Mr. Howard apparently fails to realize some very practical facts that the opponents of his theory have advanced. Among these are: the coarsely crystalline fracture of the nucleus of a transverse fissure, so different from the smooth fatigue break around it; the occurrence of fissures in large numbers in certain heats or rollings of rails, while other heats or rollings do not develop them; and the much smaller proportion of transverse fissures found in rails rolled from reheated blooms as compared with direct-rolled rails. These facts are not explained by Mr. Howard's theory unless the influence of the high-phosphorus streaks that the writer has described is admitted. The bare statement that "no structural or chemical reason common to the display of transverse fissures has been found to which their formation might be ascribed," is not a convincing reply to the evidence published, because in none of Mr. Howard's published work on transverse fissures is there any evidence as to whether the phosphorus was distributed in streaks or at the locations where the nuclei occurred.

Mr. Howard's discussion of the subject of elasticity in connection with the inception of a transverse fissure is doubtless correct in an abstract way, but it is a fact that transverse fissures develop sooner and after less internal strains in some rails than in others, and no reason is given in his argument why an uneven distribution of phosphorus in the metal ought not to be considered as a possible reason for failure under a lesser amount of strain than would be required for failure in a uniform metal. There is no question but that the relative concentration of phosphorus in one of the streaks described in my paper has a marked effect on the tensile properties of that particular fiber of metal, in spite of Mr. Howard's statement to the contrary, probably based on experience with ordinary test bars of 0.2 sq. in. (5 mm.) or larger cross-section.

The admission that transverse fissures occur "in the heavier rails after a short period of time in the track, in rails in which the direct bending stresses are much reduced," proves merely that these bending stresses

*Received Apr. 28, 1919.

are not the most important causes of transverse fissures and adds not a particle more weight to the cold-rolling-strain theory than it adds to the defective-steel-quality theory. .

Doctor Dudley's interesting discussion constitutes practically a separate paper by itself. I agree with him that there are other causes of transverse fissures beside the high-phosphorus streaks I described; segregation, for instance, was mentioned as being the cause of some of the fissures I investigated. A lag or delay in the transformations of the metal, however, cannot possibly be a cause of an internal fissure, because this phenomenon, typical of all air-cooled steel, is more effective at the surface than in the interior of the rail. I have proved this point by measurements of the relative amounts of lamellar pearlite in different parts of the sections of several transverse-fissure rails. The non-ductile cores that Doctor Dudley mentions are usually due to high-phosphorus streaks, and the enlarged grain growth, as seen in fractures, has been proved, in one case at least, to be really an instance of brittleness on account of the presence of free cementite needles, which gave the fracture a coarse appearance.

Another point that should not be accepted merely from the outward appearance is the statement that the check at the nucleus of a transverse fissure is between the grains. It has been proved that in clean metal at normal temperature the grains do not separate from each other but break through the cleavage planes. Of course the fracture at the nucleus looks different from the detail fracture around it, but more definite proof than mere looks is needed before this statement of breaking between the grains can be accepted. Another similar case is the idea that the gag makes an imprint inside the rail through ½ in. of steel; the mark on a horizontal fissure fracture may look like the imprint of the gag, but the idea that such an imprint can be transmitted through ½ in. of steel, without diffusion that would render it unrecognizable, is beyond the belief of anyone used to working with metals, and simply because a mark in a fracture looks like the imprint of the gag is not good proof that the gag made it.

Doctor Dudley's theory alone is insufficient to explain the interior origin of transverse fissures except in the few cases where ordinary segregation is present. High-phosphorus streaks as described in my paper, are one reason for this interior origin, and Mr. Howard in his theory of cold-rolling strains gives another reason to which unfortunately Doctor Dudley gives no consideration. The concluding paragraph of his discussion is no doubt true, but its connection with my paper is hardly apparent.

In spite of the above criticisms, however, I do not wish to appear unappreciative of the interesting discussion written by Doctor Dudley. The numerous results of tests and other records that he quotes to show

the better quality of rails rolled from reheated blooms are valuable as substantiating the statements to that effect in my paper, and I am much indebted to him for bringing them forward in this connection.

Mr. Davidson's questions are rather outside of my experience, but it is extremely doubtful if a thoroughly practical device will ever be developed to locate internal transverse fissures in rails in track. This seems to be a case where careful track inspection is not an effective safeguard, but preventive measures, such as not too high carbon contents, reheating of the blooms in rolling, etc., must be relied upon to give the greatest safety possible with our present knowledge.

Mr. Bronson implies that the question of the relation between the high-phosphorus streaks and transverse fissures has not been investigated enough, and questions whether these streaks are the true explanation for the superiority of rails rolled from reheated blooms. On the first point I agree with him heartily, and would welcome further investigation of the question. There can be no doubt, however, that the more even phosphorus distribution is by far the greatest structural improvement caused by reheating of the blooms in rail manufacture, and this very definite difference certainly outweighs all of Mr. Bronson's tabulated advantages, some of which are vague or of doubtful value.

The absence of transverse fissures in Bessemer rails, in spite of their high phosphorus content and streaky structure, is not fatal to the support of my theory that the streaks in open-hearth rails aid the development of transverse fissures, because these fissures are found almost exclusively in high-carbon steel without a ferrite network, which is always present in Bessemer steel. The ferrite network seems to counteract the effect of the high-phosphorus streaks, which apparently hasten the formation of fissures only in rails of practically pure eutectoid composition.

Mr. Bronson seems to be stretching his point a little when he mentions rail steel as being self-hardening in even a modified form; the "air-hardening" of the rail is most effective at the outer skin and hence does not explain the interior origin of transverse fissures. To explain this he assumes a vague interior brittleness which, although he does not seem to admit it, is in most cases merely the effect of the high-phosphorus streaks which I have described. Mr. Bronson advances no proof except fracture appearance that the nucleus of a fissure is broken suddenly by the gag, and this, as noted above, is treading on dangerous ground.

Before closing the discussion it might be of interest to describe an experiment made in our laboratory since the paper was written. Specimens were cut from four of the most badly streaked rails mentioned in the paper, at parts of the head directly adjacent to the specimens already examined. These were heated for probably about 15 hr., at temperatures around 1250° C., and after this treatment the phosphorus was found to

have diffused to such an extent that the streaks which had been distinct were now practically obliterated. This experiment contradicts the statement sometimes encountered that the laminations due to the rolling out of dendritic structure in steel are not removable by any form of heat treatment. On the contrary, it can be done, but a high temperature, such as is maintained in a bloom-reheating furnace, is necessary, and quite a long time is required to secure complete uniformity of structure.

Development of Grain Boundaries in Heat-treated Alloy Steels

BY R. S. ARCHER,* DETROIT, MICH.

(New York Meeting, February, 1919)

IN the microscopic examination of aircraft-engine parts made of heat-treated alloy steels, the writer has been forcibly impressed by the failure of the usual etching processes to disclose any but gross defects in heat treatment. It seemed particularly desirable to bring out the grain size of the steel, which is indicated in the ordinary etching processes chiefly by the change in orientation of structure from grain to grain. After various attempts, a fairly successful method has been found, which is here presented in the hope that it may be of use to those who have occasion to examine these steels in the heat-treated condition. The specimens worked with have been mostly chrome-nickel steels heat treated (that is, hardened and tempered) to a Brinell hardness of around 300. A few experiments with other specimens will be referred to later.

Briefly, the method employed consists in etching the steel about 10 min. in a fresh solution of picric acid in ethyl alcohol, and then rubbing off the smudge of carbonaceous matter on moist broadcloth. Different samples of almost identical analysis, heat treatment, and physical properties (including resistance to impact) respond quite differently to this treatment. In some, the grain boundaries are developed quite readily, while in others, especially in those whose grain is unusually fine or in which the final hardening heat has not quite obliterated the previous crystallization, the results are more obscure. The exact time of etching to secure the best results also differs for various samples, and the greatest success is usually attained by a process of "tinkering." Such a process is illustrated in a series of micrographs, the first six of which were taken at a magnification of 300 diameters.

The properties of the steel are as follows:

Carbon, per cent	0.42	Brinell hardness	293
Nickel, per cent	1.35	Yield point, pounds per square inch	131,000
Chromium, per cent	0.90		
Copper, per cent	0.025	Maximum strength, pounds per square inch	144,500
Manganese, per cent	0.74		
Phosphorus, per cent	0.024	Elongation in 1 in., per cent	28
Sulfur, per cent	0.036	Izod impact, foot-pounds	51
		Reduction of area, per cent	57.3

*Bureau of Aircraft Production

The specimen was etched for 2 min. in a fresh solution of 5 gm. of picric acid in 100 c.c. of denatured alcohol. It was then rubbed gently in all directions on moist kersey cloth, the fine polishing wheel being used for this purpose. Its appearance after this treatment is shown in Fig. 1. The specimen was then etched 2 min. more in the same solution and again wiped on cloth. After photographing, this process was repeated up to a total of 15 min. The results are shown in the micrographs (Figs. 1 to 5), the total time in the etching solution being given below the micrographs.

By this process the clearest development of grain boundaries was obtained at the end of 11 min., the result after 4 min. more of etching being inferior. The sample was then repolished and etched again for 11 min. continuously in a fresh solution. The result was better than before, and is shown in Fig. 6. Fig. 7 shows this structure at a magnification of 1000 diameters. These photographs do not show the same area as is shown in the first series, since the surface layer has been changed in the repolishing. Fig. 8 shows the appearance of the same specimen as etched for 2 min. in the same solution and dried without any rubbing.

With the same specimen of steel, a solution of 5 gm. of picric acid to 100 c.c. of 95 per cent. ethyl alcohol gave results practically identical with those obtained with the solution in denatured alcohol. Absolute alcohol was not tried, nor was any dehydrating agent, such as acetic anhydride. Alcoholic solutions of o-nitro-phenol were found to act too slowly. Mixtures of this reagent with the alcoholic solution of picric acid acted with correspondingly less speed than picric acid, and gave results which were possibly a little more uniform, but not of sufficient improvement to justify the extra time. The same effect might be obtained by diluting the picric-acid solution with alcohol, but this was not tried. The following reagents were tried unsuccessfully: nitric acid in ethyl alcohol, concentrated nitric acid, bromine in alcohol, and bromine in carbon tetrachloride.

A few experiments have been made with other steels and with chrome-nickel steels of different heat treatment. Attempts to develop the grain boundaries in specimens from chrome-nickel gears, made of S.A.E. steel X 3345 and tempered to a Brinell hardness of about 500, failed to give similar results to those obtained with the same steel in a softer condition. A straight carbon steel, of medium carbon content, was quenched in water from 1600° F. (872° C.) and drawn at 1050° F. (566° C.). The grain boundaries could not be brought out by this method. Attempts with a chrome-vanadium steel were at least partly successful. The steel in the shape of $\frac{5}{8}$-in. (15.9-mm.) round bars was quenched in water from 1700° F. (922° C.) and drawn at 1050° F., and had the following properties.

FIG. 1.—2 MIN.

FIG. 2.—4 MIN.

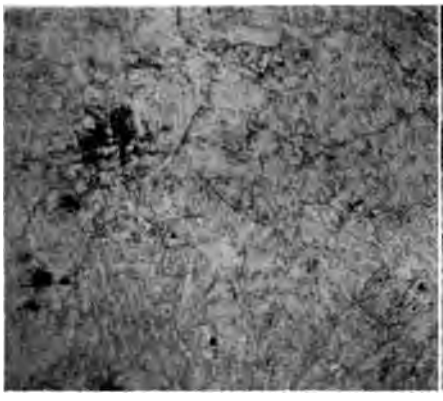


FIG. 3.—7 MIN.

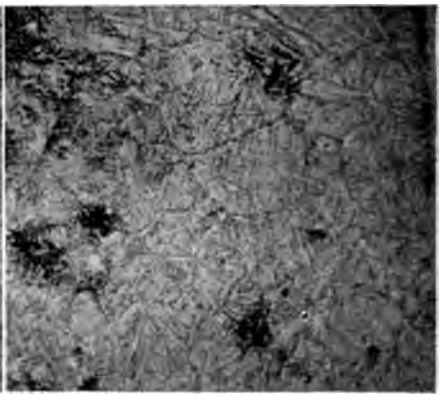


FIG. 4.—9 MIN.

FIG. 5.—11 MIN.

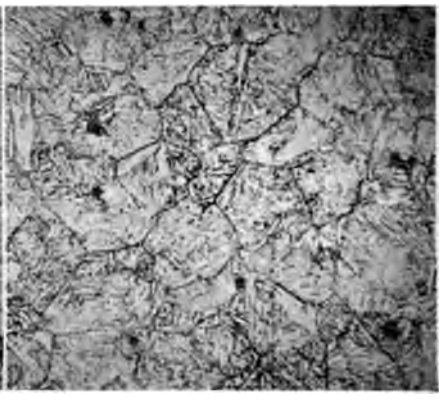


FIG. 6.—11 MIN. CONTINUOUS ETCHING.

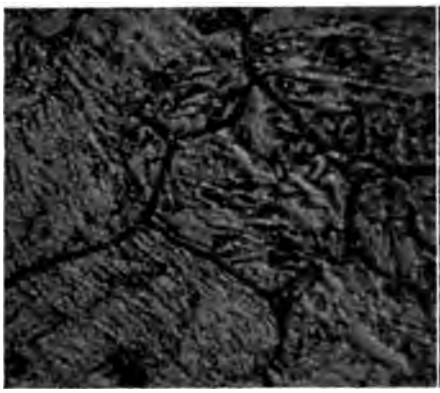


FIG. 7.—SAME AS FIG. 6. × 1000.

FIG. 8.—ETCHED 2 MIN. WITHOUT RUBBING. × 300.

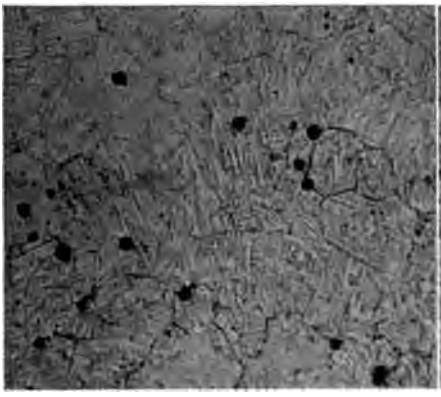


FIG. 9.—CHROME VANADIUM × 300.

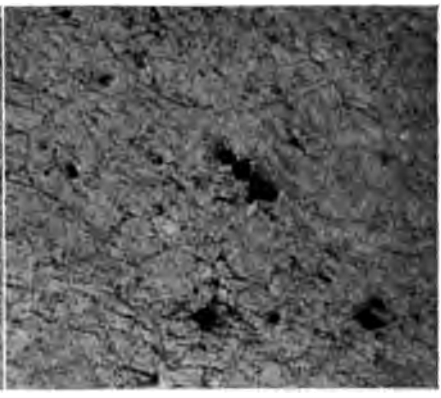


FIG. 10.—CHROME VANADIUM, ETCHED WITHOUT RUBBING. × 300.

FIG. 11.—PATH OF RUPTURE IN BRITTLE STEEL. × 250.

FIG. 12.—INCIPIENT FAILURE IN BRITTLE STEEL. × 250.

Carbon, per cent	0.27	Brinell hardness	302
Chromium, per cent	0.96	Yield point, pounds per square inch	146,000
Vanadium, per cent	0.21		
Manganese, per cent	0.57	Maximum strength, pounds per square inch	152,100
Phosphorus, per cent	0.023		
Sulfur, per cent	0.031	Izod impact, foot-pounds	52
		Elongation in 2 in., per cent	16.9
		Reduction of area, per cent	60.6

Fig. 9 shows this steel etched 20 min. in picric acid and rubbed, while Fig. 10 shows the result of 2 min. etching without any rubbing. It may be remarked that often a development of grain boundaries can be obtained which makes it possible to get a good idea of grain size by visual examination of various fields of the specimen, but which is not good enough for an accurate grain count in any one area. The grain size is usually such as to render the 4-mm. objective convenient for observation and photography. The writer's brief experience with this method has shown that in some brittle steels the grain outlines are developed more easily than in tough steels of similar analysis and tensile properties.

The method has been tried for tracing the path of rupture in impact test specimens. It works satisfactorily in the case of very brittle steel, where the fracture is shown to take place around the grain boundaries; Fig. 11 shows such a specimen. In Fig. 12, taken from the same specimen, it can be seen that failure has started, in the grain boundaries, below the actual surface of fracture. The properties of this steel were:

Carbon, per cent	0.49	Brinell hardness	302.
Nickel, per cent	1.90	Izod impact test across grain, foot-pounds	8.5
Chromium, per cent	1.02		
Manganese, per cent	0.74	Izod impact test parallel to grain, foot-pounds	2.0
Sulfur, per cent	0.037		
Phosphorus, per cent	0.037		

When the resistance to impact is high, the fracture is through the grains. The resulting deformation of the grains causes failure of the etching process to bring out the grain boundaries near the fracture. The manner of failure, however, is readily apparent from the profile of the fracture.

Summary

A method of etching is described which has been found useful in bringing out the grain boundaries of chrome-nickel and chrome-vanadium steels hardened and tempered to a Brinell hardness around 300. This method consists in etching the specimen in a 4 per cent. solution of picric acid in ethyl alcohol for a period varying from 5 to 25 min., and then rubbing off the carbonaceous smudge on moist broadcloth or kersey.

DISCUSSION

J. A. MATHEWS,* Syracuse, N. Y. (written discussion†).—The reagent described by Mr. Archer seems to be a selective reagent, useful in one particular field. Nevertheless, the field is one in which it is likely to prove of considerable value, because there is a great deal of alloy steel heat treated to approximately the properties mentioned in the paper. If it is found, by further tests, to be applicable to a wider range of compositions than merely chrome-nickel steels, for the most part discussed by Mr. Archer, its value will be thereby increased to just the extent to which it can be applied. In view of the fact that certain of these chrome-nickel steels, very similar in physical properties and hardness, show very

FIG. 13.—ETCHED 1 HR. IN ARCHER'S REAGENT. IZOD 60. × 188.

FIG. 14.—ETCHED 1 HR. IN ARCHER'S REAGENT. IZOD 10. × 188.

different results in the Izod or notch test, and in view of the fact that when such differences are noted in the fracture the brittle steels usually manifest the appearance of intercrystalline failure while the tough ones show intracrystalline breaks, it occurred to me that this etching reagent might in itself give advance information as to tough and brittle steels.

Through the courtesy of Mr. J. H. Nelson, I was furnished with a series of steels covering a range from extreme brittleness to extreme toughness. In the etching experiments we gave considerably longer time than that noted by Mr. Archer; trying them, first, for ½ hr., and later increasing

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† Received Feb. 17, 1919.

this time to a full hour. Under these conditions we developed very beautiful grain boundaries, but there is nothing to indicate that the brittle steels were more or less easily attacked than the tough ones. It was, however, immediately noticed that the tough steels were very much finer grained than the brittle ones. We made an actual grain count upon the series of steels and in the accompanying table we give the correspond-

Comparison of Izod Figures and Grain Size

Izod Figure	Number of Grains per Sq. Mm.
60	2103
54	2762
37	1527
4	914
10	875
9	1256

ing Izod figure and the number of grains per square millimeter. Two photomicrographs from the series of six tests made illustrate the marked difference. It will be observed that there is not a direct relation between the two sets of figures, which could hardly be expected. The steels in question are of the same kind as the chrome-nickel steels discussed by Mr. Archer.

Blast-furnace Refractories

BY RAYMOND M. HOWE, PITTSBURGH, PA.

APPENDIX *

Additional data have been secured on the disintegration of furnace linings and the spalling of stove brick.

The first article on the disintegration of furnace linings appears to have been written by John Pattinson in 1876.[1] The furnace involved was 85 ft. (25.9 m.) higher and 27 ft. (8.2 m.) in diameter at the boshes; after being in blast five years, the top brick became contaminated with carbon and disintegrated. An examination of the lining showed that the discolored, disintegrated brick were in a zone found from 35 to 67 ft. (10.6 to 20.4 m.) above the hearth; immediately below, in a zone extending from 22 to 40 ft. (6.7 to 12.1 m.) above the hearth, the brick were hard and porcelain-like in appearance. Analyses of samples taken from the different sections are given in Table 1. The water-soluble portion found in the sample taken 39 ft. (11.8 m.) above the hearth had a composition approximately as follows: silica, 0.50 per cent.; lime, 0.04 per cent.; potash, 0.78 per cent.; soda, 0.20 per cent.; potassium chloride, 6.19 per cent.; moisture, 1.03 per cent.; total, 8.74 per cent. The carbon

TABLE 1.—*Analyses of Samples Taken from Different Sections of Furnace*

	I Original Lining Per Cent.	II Section 67 ft. Above Hearth Per Cent.	III Section 45 ft. above Hearth Per Cent.	IV Section 39 ft. above Hearth Per Cent.	V Section of Vitrified Portion Per Cent.
Silica	55.70	55.60	55.67	47.60	37.07
Alumina	35.50	35.40	30.00	26.47	23.18
Ferric oxide	4.00	2.67	3.00	1.32	0.96
Lime	0.30	0.34	0.23	0.06	0.17
Magnesia	1.21	1.22	0.95	0.86	0.00
Potash	2.60	2.10	5.30	7.54	21.59
Soda	0.45	0.42	1.67	0.76	6.28
Carbon	....	2.05	3.24	5.97	
Carbon dioxide	....	....	....	....	2.17
Water	....	....	....	....	7.70
Water-soluble	....	....	....	8.74	

* See pages 1–17, this volume.

[1] On Carbon and Other Deposits from the Gases of Blast Furnaces in Cleveland. *Jnl.* Iron and Steel Inst. (1876), 85.

deposits consisted of 84.9 per cent. carbon and 15.1 per cent. ash, most of which was oxide of iron.

Judging from Pattinson's statements the conditions in the furnace might be analyzed as follows: The charge contained appreciable amounts of alkali (probably as chlorides) which volatilized in the vicinity of the bosh. These chlorides either permeated the lining in the hotter portions, condensed on the cool portions, or passed out of the furnace. Where the temperature conditions were favorable, the chlorides behaved in the usual way and reacted with the iron to form potassium oxide, sodium oxide, and volatile ferric chloride. The soda and potash then reacted with the fireclay to form a dense porcelain-like body having a composition similar to that of true porcelain. The slightly cooler lining above this was contaminated with chlorides which caused some of the iron to volatilize but which did not react to form a dense hard structure. It was in this zone that the carbon deposition was greatest (see analysis IV). The still cooler sections above this were not even penetrated by the alkalies.

Naturally Pattinson concluded that the skin, or coat, which extended to about 40 ft. (12 m.) above the hearth, protected this part of the lining from the various destructive influences. He also concluded that a low iron content in firebricks is most desirable, for the carbon deposits build up about iron spots. Furthermore, he suggested the use of very hard, dense, top brick as a means of duplicating the impervious coat which he believed to be so favorable to long service.

After finishing his laboratory work, he visited several other plants in order to see whether or not the condition noted by him was rare or one that had been overlooked. He found a distorted flue lining, a disintegrated furnace lining, and a section near a burst shell, all of which showed excessive carbon deposition at or near the point of failure.

I. L. Bell, M. P., discussed this paper and stated that iron causes a deposition of carbon from flue gas. Successive layers are built up from the surface of the iron and exert a tremendous disintegrating or disrupting effect upon the containers.

The chemistry of this phenomenon remained obscure for several years after Bell and Pattinson had made their findings public. In 1891, however, Messrs. Mond and Quincke prepared iron carbonyl by passing carbon monoxide over iron for a period of six weeks at a temperature of from 200 to 350° C.[2] This combination between iron and carbon monoxide was observed to be sluggish and extremely sensitive to temperature changes. When higher or lower temperatures were involved, metallic iron, carbon and carbon dioxide were formed. Thus the penetration of

[2] Note on a Volatile Compound of Iron and Carbonic Oxide. *Jnl.* Chem. Soc. *Trans.* (1891) **59**, 604.

iron by carbon and the precipitation of carbon from carbon monoxide appear to be similar to the penetration of silver by oxygen and of platinum by hydrogen.

This reaction was further studied by Schenck and Zimmerman, in 1905.[3] They found that the deposition of carbon from carbon monoxide did not take place in the presence of iron oxide but that the presence of iron or certain other metals was necessary. When such a metal was oxidized by CO it took place in two steps.

$$Fe + 2CO = Fe + CO_2 + C \quad (I)$$

$$Fe + CO_2 = FeO + CO \quad (II)$$

They also found that if the partial pressures of carbon monoxide and carbon dioxide in the furnace, taken together, are greater than the critical pressure, the oxidation of iron takes place with the deposition of carbon; if the pressure is less, carbon can be deposited but no iron will be oxidized. The critical pressure varies only with the temperature and is as follows:

500° C	15 mm.	700° C	305 mm.
550° C	35 mm.	750° C	535 mm.
600° C	70 mm.	800° C	800 mm.
650° C	145 mm.		

F. W. Lurmann, in 1898,[4] showed how all refractory clays contain pyrites, which burn to form iron spots. These spots are the centers for carbon deposition, which takes place at from 300 to 400° C. He had observed that firebricks (made at the same time) remained strong when used at higher or lower temperatures, while those subjected to blast-furnace service at the indicated temperatures disintegrated.

TABLE 2.—*Analysis of Original Brick, Coating, and Protected Brick*

	Original Brick	Coating	Brick 1 inch Inside of Coating
SiO_2	55.62	40.23	57.63
Al_2O_3	38.55	12.22	31.64
Fe_2O_3	4.17	11.93	3.73
CaO	0.24	10.92	0.24
MgO	0.24	4.31	0.11
K_2O	0.95	9.39	2.59
Na_2O	0.29	8.43	0.51
Zn	.....	.....	1.51
ZnO	.....	.....	0.59

[3] *Stahl u. Eisen* (Oct. 1, 1905) **25**, 1121.

[4] *Stahl u. Eisen* (1898) **18**, 168.

Firmstone described a more recent failure.[5] He found that a good coating had protected the brick up to a height of 40 ft. from the bottom, while above this the bricks had disappeared entirely. An analysis of this coating, of the original brick, and of the protected brick is given in Table 2.

Firmstone did not discover the presence of soluble salts in any of the remaining lining nor did he discover appreciable amounts of carbon. He did find that the bell was impregnated with carbon and that the protective coating was high in alkali. Therefore, he concluded that the conditions under consideration were duplications of those observed by Pattinson. He stated that the furnace had been stopped several times because of the shortage of coke and that it was badly scaffolded. These scaffolds were responsible for hot and cold spots resulting in producing a wide temperature range, part of which would favor carbon deposition.

Bernhard Osann[6] does not accept the carbon deposition theory of disintegration but intimates that this is due to the presence of alkalies. These various opinions might be summarized as follows:

Iron (and possibly its oxides) will cause a precipitation of carbon from carbon monoxide at temperatures between 300° and 400° C. This action, although slow, may result in the building up of large carbon deposits, about relatively small centers, and in so doing disrupt the material. The necessary conditions for such deposition and subsequent disintegration prevail in blast-furnace tops. Since carbon deposition has been heavy in several of the linings which failed, many authors believe that this is responsible for their failure. Furthermore, they believe that this condition is more prevalent than is generally appreciated.

Alkali chlorides are present in many furnaces and penetrate the lining. They react with the iron present, forming iron chloride, which is volatile. When the temperature is sufficiently high, these chlorides combine with the lining and produce a strong, impervious, porcelain-like protective coating. This coating protects the inner lining from the action of furnace gases, the erosion of the charge, and similar destructive influences.

The writer has had but one experience with such disintegration and the causes for this were so obscure that no definite conclusions could be drawn. Very little carbon was found deposited in the disintegrated firebrick, which were of at least average quality, were strong, well burned, and of the necessary refractoriness. This latter fact was brought out by the usual load and "fusion" tests, which showed the top and inwall brick to soften at cone 31 and the hearth and bosh brick at cone 33; neither compressed abnormally in the load test. In view of these conclusions, the writer concluded that, if the deposition of carbon had

[5] Alteration of Firebrick by Furnace Gases. *Trans.* (1904), **34**, 427.

[6] *Stahl u. Eisen* (1903) **23**, 823.

disrupted the firebrick, a later increase in temperature had caused it to burn out.

DISCUSSION

There are two important points that should not be overlooked, one of recent observation and the other of a purely hypothetical nature. The first point concerns the disintegration of glass pots by the action of hydrogen chloride at red heat. The conditions for the formation of this compound are present in the blast furnace and its action may be vital. It is quite possible that Osann was correct in believing that alkalies, introduced as chlorides, caused disintegration for the formation of hydrogen chloride through the interaction of alkalic chlorides and steam would introduce a very active gas.

1. The use of a top lining which is absolutely free from iron.
2. The use of a minimum furnace temperature of over 400° C.
3. The installation of impervious blocks in the top and middle inwall.

The first solution is entirely impossible, for the best porcelains contain iron. If the action of this element is as great as some authorities claim, the slightest trace would be capable of doing untold damage, under favorable conditions.

The second possibility, likewise, is impractical, for even if the inside minimum temperature were 400° C. or more, there would be inner zones according to the temperature gradient.

The third solution is more in accordance with plant observation. In the different cases under discussion, each writer associated a protective coating with satisfactory service, for disintegration took place above the upper limit of this coat. However, this condition may have been simply a combination of circumstances, for it requires rather high temperatures to form such a coating while low temperatures favor carbon deposition.

In spite of this question as to the role of these coatings, there is no doubt that their presence is desirable and that they should extend the entire length of the lining. The formation of such in the top of the furnace during service is impossible, however, for temperatures of at least 900° C. are required to bring about the necesssary reactions. Therefore, it would be necessary to install such a protective coating in the same way as the usual lining. It might be installed in the form of blocks consisting of feldspar, silica, and fireclay, with the possible addition of sodium silicate. Such a mixture would possess the necessary refractoriness to meet top and possibly inwall requirements. In case it were not sufficiently refractory it could do no more than soften and glaze the lining, thus accomplishing the object at hand. It would be practically impervious to gases and should be resistant to abrasion and impacts.

Spalling of Stove Brick

Mention has been made of the fact that considerable spalling had been observed in the lining of blast-furnace stove combustion chambers and in the top courses of checkers. It was the writer's privilege to secure samples of firebrick that had spalled and also of the spalls. The general characteristics of these firebrick were found to be as follows: They are made on an auger machine, from a highly refractory plastic clay which fused at Orton cone 31. These bricks show a compression of about 8.33 per cent. in the standard 1350° C., 25 lb. per sq. in. load test. They lose about 15 per cent. of their weight when heated twice to 1350° C. for a period of one hour and plunged into cold water. The general analysis of these brick is given in Table 3, together with analyses of two spalls coming from them.

Table 3.—*Analysis of Stove Brick and Spalls*

	Original Firebrick	Spall No. 1	Spall No. 2
Silica	59.48	59.02	44.96
Alumina	35.92	36.10	28.10
Ferric oxide	2.02	2.10	11.40
Lime	0.30	0.26	3.40
Magnesia	0.21	0.28	0.98
Alkalies	1.86	2.01	10.30

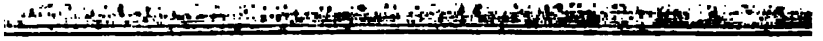
The fusion and load tests show that these bricks are similar in refractoriness to top and inwall brick. The dipping tests show that their structure is not particularly adapted to resisting the more severe conditions of temperature change. The analyses of the spalls show that in one case (No. 2) the tendency to spall had been accelerated by the deposition of an easily fused dust, high in iron and alkalies. This dust had apparently melted and solidified alternately, thus weakening the structure of the brick. Analysis fails to indicate any such cause for the spalling of the other sample.

The combined results of the laboratory tests and plant experience lead to the following conclusions: (1) A firebrick of the indicated characteristics is entirely satisfactory for most of the stove checkerwork, but is not entirely satisfactory for the lining of the combustion chamber and the top checkers. (2) The primary cause of failure may lie in the firebrick itself, for it may not have the necessary structure for resisting severe conditions of temperature changes; the influence of these temperature changes is also accelerated by the deposition of dust.

The following results were secured from tests made on similar firebrick and undoubtedly apply to the case at hand, for the clays involved are much the same. Therefore it is thought that the incorporation of, say, 30 per cent. flint clay or ground bats in the mix will overcome the spalling trouble to a large degree.

Batch	Number of Dips in Spalling Test Causing Failure
100 per cent. plastic clay	3
90 per cent. plastic + 10 per cent. flint clay	4
80 per cent. plastic + 20 per cent. flint clay	3(?)
70 per cent. plastic + 30 per cent. flint clay	6
60 per cent. plastic + 40 per cent. flint clay	over 22

BIOGRAPHICAL NOTICES



Andrew Carnegie

ANDREW CARNEGIE, America's best-known ironmaster and philanthropist, died at his home at Lenox, Mass., on Aug. 11, 1919, after a three days' illness. A pioneer in the steel industry, he introduced into this country the Bessemer process and formed the plans that have resulted in this country's supremacy in the steel markets of the world. Known throughout the world for his gifts, he is also known to engineers as the donor of the home of their National Societies and as the founder of the Carnegie Institute of Technology, of Pittsburgh. For many years, also, he was an honorary member of the American Institute of Mining and Metallurgical Engineers, the American Society of Mechanical Engineers, and of the Engineers Club of New York. What he considered the secret of his success is shown by the epitaph he wrote for himself, "Here lies a man who knew how to enlist in his service better men than himself." He was buried Thursday, Aug. 14, in Sleepy Hollow Cemetery, Tarrytown, N. Y., the funeral services being private, in accordance with his wish.

TRIBUTE OF C. M. SCHWAB AND JAMES GAYLEY

It would be difficult for me to find words to express my love and admiration for Mr. Carnegie, my friend, partner and associate for forty years. He was the greatest man I ever knew, and he had a heart so filled with tender sentiment, especially with reference to his associates, as to make him beloved as well as admired by all those who came in business or social contact with him. He possessed the faculty of inspiring others to unusual efforts in a greater measure than any man I ever knew, and he always won by expressions of appreciation rather than criticism. The world has lost a great man and a great benefactor to humanity, and I have lost a greater friend than whom no man ever had.—C. M. SCHWAB.

Mr. Carnegie was unquestionably the foremost manufacturer of steel in the world, and was a force that influenced the economic life of nearly every country. This achievement seems the more remarkable for a man who once regretted that he had ever quit the railroad business, and we can well imagine what he would have achieved in the world of railroads, if he had made it his life long work.

He was a man of unbounded energy, and once told me that prior to an attack of typhoid fever (at about fifty years of age), he had never

known what it was to be physically tired. Mr. Carnegie was never hampered by traditions in the steel industry, but strode out in bold and original lines. He feared no rivals. He was always ready to put back earnings into the plants to effect economies, and was foremost in expanding the mechanical age in steel production. He quickly foresaw that wooden bridges for railroads were soon to pass, and established a works to build bridges of steel. The necessity for cheaper and better rails arising, he early installed the Bessemer process and was a pioneer in its development. Wooden cars, like wooden bridges were no longer adequate to the demands for cheap transportation, so cars of steel with greater load capacity were constructed, and although these were condemned by the engineers of a great railroad, he unhesitatingly equipped the Bessemer and Lake Erie R. R., which he had then recently acquired, with steel cars to carry the ore from Lake Erie to the Pittsburgh works. Today steel cars are the common bulk carriers on all great freight systems.

In many ways, he popularized the use of steel, as for example, in the substitution of soft steel for wrought iron, and demonstrators were sent out even to blacksmith shops to show the superiority of steel.

When Mr. Carnegie first entered the iron industry, things were conducted in a haphazard way. He found it extremely difficult to learn from other manufacturers the amount of coal required to produce a ton of puddled iron, so he inaugurated a careful system of works practice and costs to work intelligently.

He claimed no acquaintance with mechanical or metallurgical details, but through his daily, weekly and monthly reports he kept his fingers on the keyboard of operations, and no one was more familiar with the cost sheets than he. He loved the field of big things, of big tonnages, and left the specialties to other manufacturers.

He has frequently stated that his greatest success in life came from his ability to select men competent to manage his affairs. Seldom did he go outside his own organization to choose a manager. He preferred men acquainted with the Carnegie methods, and rarely was he mistaken in his choice. To each head of a department, he gave his full confidence and coöperation. From the managers at the works, successful operation was demanded, but there was no hesitation in authorizing every expenditure, great or small, that the manager could demonstrate as contributing to greater outputs or economics. To every man in his employ, there was the opportunity for advancement—those who made good were taken into the partnership, and thus there gradually grew an organization of great efficiency, stimulated to a degree that would never have been possible through large salaries, united in a common interest, where jealousies and hours of service were forgot, and business became a pleasure; and above all was the vital inspiring force the presiding genius of Andrew Carnegie.

Mr. Schwab has well said that Mr. Carnegie stimulated by expressions of appreciation and not through criticism. He never condemned experiments which failed, but encouraged the manager by claiming that it was just as important to know what not to do.

With his clear vision, he often foresaw things better than his chief officials, and dictated the policy, but he never went contrary to a unanimous vote of the Board of Directors.

He did business in the open, and was never afraid to show his hand, and in all matters was particularly frank with his associates. Contrary to the usual custom of the day in this country and abroad, his works were always open to visitors, and it became clear that the casual visitor could not gage the factors that made for success.

Mr. Carnegie's day was that of the great expansion of the iron and steel industry, and in the age when metallurgical and mechanical methods made its greatest stride his name will ever live. He was an organizer, manufacturer and salesman combined, and never lost faith in a bright future for the steel industry.

In 1885 when he was recuperating from typhoid fever at Cresson, Pa., he said to his great steel manager, Captain W. R. Jones, that during his serious illness, he always wished to live to show what a man could achieve in a lifetime. The achievements of Andrew Carnegie are ended; that man of genius has passed away, and it is now left to the world to place an estimate on those achievements. JAMES GAYLEY.

BRIEF BIOGRAPHY *

The good fortune of Andrew Carnegie began at the very start. He was born in Dunfermline, Fifeshire, Scotland, on November 25, 1835. His good fortune in the sense conceived lay not in the place but the year of his birth—a year that synchronized with certain conditions in America which left him just about time to qualify for the part he was to play in the story of steel. He needed thirty years for the intermediate steps, and he had them, with one to spare, before the high tariff had thrown its protecting arms about the "infant industry" he was to help toward maturity. The tariff and control of transportation lay at the very foundation of his success.

How he came to command that golden rebate privilege and convert to one gigantic purpose the Bessemer steel process, Lake Superior ore and Connellsville coke is explained by some of the earlier steps in his restless progress upward.

There was but little education for the boy in Dunfermline. What little he got was at the local schools and from an uncle who was addicted to inflammatory speech and labor agitation. In 1848, the general adoption in Scotland of the steam loom found his father, a weaver, out

* Reprinted from New York *Evening Sun*, Aug. 11, 1919.

of employment and the family decided to emigrate to the United States. After a voyage that lasted 49 days they reached port, mother and father, Andy, 13 years old, and Tom, 6 years old. They made their first home in Barefoot Square, Slabtown, Pa.

Here the father found work at his trade and entered Andrew in the same employ as a bobbin boy at $1.20 a week wages. The mother took in washing and did binding for a neighbor, a shoemaker named Phipps. With the shoemaker's 10-year-old son Harry young Andrew earned odd pennies peddling fruits and cemented a friendship and later a partnership not to be completely dissolved while both lived.

Relations with Colonel Scott

The duties of bobbin boy were not sufficiently remunerative and Andrew resigned. After a brief essay at tending furnace he became a messenger boy in the employ of the Ohio Telegraph Company at a salary of $11.25 a month. In his resting moments he studied telegraphy, and 4 years later obtained employment with the Pennsylvania Railroad as an operator.

Willingness to take a chance, with the ability to back it up, resulted in his next advancement. A wreck occurred on the railroad and Colonel Scott, who was in charge, was not to be found. Young Carnegie realized the conditions and rose to the occasion. He sent telegrams broadcast to train despatchers at other stations ordering trains held and signed them "Thomas A. Scott, Division Supt." Traffic was restored without further accident. For this notable service he was made secretary to Colonel Scott.

As secretary to Colonel Scott, he found himself bound up in a number of enterprises and turning his promiscuous activity to profit. It started with an opportunity thrown his way by Colonel Scott to purchase ten shares in the Adams Express Co. at $60 a share. His mother mortgaged her house to raise the money and he bought the stock. The transaction opened to him a vast perspective of possibilities. He had become a capitalist.

Before long, in one way and another, he had bought his way into the Woodruff Sleeping Car Co., the Columbia Oil Co., the Duck Creek Oil Co., the Dutton Oil Co., the Pittsburgh Elevator Co., the Third National Bank of Pittsburgh, and a locomotive and bridge works. The 10 years from 1855 to 1865 were devoted to these interests without the surrender of his Pennsylvania connection. At 28 years of age, in 1865, he succeeded Colonel Scott as superintendent of the Pittsburgh division.

An oil speculation yielded him his first large profit. The transaction involved the outlay of not a single dollar. A note bought the stock; the dividends paid for it.

Early Iron Ventures

In 1864, Carnegie began to appreciate the possibilities in iron. He bought a one-sixth interest in the Iron City Forge Co. Thomas N. Miller, Andrew Kloman, and Henry Phipps were the other partners; Miller was voted out and induced Carnegie to join with him in starting rival works. The experiment was not a success and the two companies were merged as the Union Iron Mills Co. with a capital of $500,000. Carnegie then organized another company, the Keystone Bridge Works, which was prosperous from the first. Into it he drew Colonel Scott, vice-president of the Pennsylvania Railroad, and J. Edgar Thomson, its president. The dividends of the new company quickly showed how much it was worth to have the support of a great railroad. Carnegie resigned as superintendent. He had formed a more profitable connection with the railroad.

At the outbreak of the Civil War, when Colonel Scott was made Secretary of War, Carnegie acted as superintendent of military roads. The knowledge he then gained enabled him later to bid advantageously for many contracts for new bridges and iron and steel for new railroads.

In 1867, he represented the Pennsylvania Railroad in England, being commissioned to sell about $9,000,000 in bonds. The task was fraught with difficulties, but Mr. Carnegie accomplished it, and his commissions amounted to approximately $225,000. Discharging his trust with the railroad, he kept a strict watch on the steel industry in England, and in 1868 returned to the United States and installed the Bessemer process of steel refining in his mills.

Results of Adoption of Bessemer Process

It was the adoption of the Bessemer process which laid the way open to empire in steel making. While abroad Carnegie saw a giant Bessemer converter in operation and investigated the process. From that moment his vision was steel. He returned to Pittsburgh and organized the firm of Carnegie, McCandless & Co. There were eventually fourteen partners in the concern. Into its coffers went all Carnegie's profits from the sale of bonds and $25,000 more.

The first step was the purchase of a 110 acre tract a dozen miles or so outside of Pittsburgh, where they erected a Bessemer plant and named it the Edgar Thomson Steel Works. With Thomson, president of the Pennsylvania, there was gold in the circumstances. At a stroke it settled the question how to deal with competition. It was to be rebated out of existence.

Thus was described the division of labor among the new partners at this period: "Shinn bossed the show, McCandless lent it dignity and

standing, Phipps took in the pennies at the gate and kept the payroll down, Tom Carnegie kept everybody in a good humor, and Andy looked after the advertising and drove the band wagon."

Carnegie's business methods earned him no little criticism, but in results they justified themselves. He never pretended to be a practical master of the iron and steel trade. He was a student of opportunity and a salesman.

In 1874, the real Carnegie boom began. The year before, the first of the famous Lucy furnaces had been built, to be duplicated two years later. They forced up the average output of iron from 50 to 100 tons a day. The Edgar Thomson Steel Works were running a successful race with their nearest competitor.

As the profits mounted, one by one Mr. Carnegie's partners dropped out. Democracy in business had at that time no virtues for the man who was gradually emerging as master. He saw that in the clash of many opinions there was weakness; in one-man power strength and individual profits. Some of the partners are said to have been forced out; others voluntarily parted with their holdings; others died. By 1881 Carnegie owned more than half of the entire business. A reorganization was undertaken under the name of Carnegie Brothers & Co.

The new concern was a $5,000,000 affair with Carnegie at the head. It operated the Edgar Thomson Steel Works, while a second limited partnership called Carnegie, Phipps & Co. operated the Homestead Mills, the armor plate mill near the same plant, the Keystone Bridge Works and other properties. A few years later, awaiting the advent of Henry Clay Frick in 1888, Phipps and Carnegie alone remained of the original fourteen partners. In the last-named year Carnegie found himself in control of seven great iron and steel works and possessed of some $15,000,000. The profits of the companies in 1899 were given as $21,000,000.

Value of 10 Years Efficiency

During the 10-year period Mr. Carnegie had risen to be absolute master of steel. In 1892 occurred the Homestead strike and the famous Frick-Carnegie fight. Carnegie was severely criticized for having remained in Scotland playing golf while the strike was in progress, leaving Frick here to battle for them both, and for having later forced Frick out of the combine. The differences of the two men were ultimately adjusted by a plan of reorganization which recognized Frick's claims.

For 2 years from 1899 the Carnegie Steel Co. dominated the steel industry, but in 1901 it was absorbed by the United States Steel Corpn. for $460,000,000, the greatest sale ever recorded. That same year Mr. Carnegie retired from business; at that time his income was said to exceed $22,000,000 a year.

Mr. Carnegie was one of the world's efficient little men who gain in vitality what they lose in bulk. He was hardly more than 5 ft. 4 in. in height. His feet and hands were small, his eyes blue and keen, his head rather large and solidly modelled. A short white beard hid evidence of obstinacy. A shaggy shock of white hair denied his years. In his brighter moods he was a jovial figure of a man, with something of the familiar "Andy" showing through the master of millions. He was given to boyish humors and boyish self-satisfaction with what he was and had done.

[Establishes Homes in New York and Scotland]

Somewhat intolerant in his earlier years, he is said to have softened considerably after his marriage. At 53, he married Miss Louise Whitfield, the daughter of Mrs. Frances D. Whitfield of 35 West Forty-eighth street, New York. His bride was 28 years old. Their first New York home was in East Fifty-first street, but, deciding to establish his residence in this city, he built a palatial home further up Fifth avenue, extending from Ninetieth to Ninety-first street. Nothing was overlooked in this masterpiece of the architect's art to provide for the comforts and luxuries of its wealthy master.

While this home was being prepared for him, he negotiated the purchase of Skibo Castle, in Scotland. This ancient baronial mansion he completely remodelled, shipping the steel and structural iron from his mills in Pittsburgh to the shores of Scotland. Every whim of Mr. Carnegie's was gratified in the building and reconstruction of the Skibo estate. The property extends for miles along the coast. It has many excellent hunting coverts, trout streams, and a golf course laid out according to the latest expert advice. At a convenient point is a specially built iron and concrete pier, off which his private yacht was moored, always ready for sea. Over the Skibo estate, comprising some 35,000 acres in all, flew a special flag made to show a combination of the Stars and Stripes with the Union Jack.

In this country retreat the tenants of Skibo knew a different side of the venerable owner than did the world. In business, he was hard, sharp, and aggressive, demanding the same full value he gave. In Skibo, he was the generous dispenser of gifts to the people. He was a promoter of games and took keenest pleasure in watching the sports. A private piper in Highland costume accompanied him on all his trips about the estate. Nothing seemed to give him more delight than watching the foot races of the boys and girls of the place. Lacking an appreciation of amateur sporting laws, he gave liberal cash prizes.

He offered many advantages in education to the children of Skibo. He remembered his own childhood, with its bitter struggle against the wolf, and always sought to ease the path of the kiddies who came to

his attention. The upkeep of such a tremendous tract required the maintenance of a large staff of assistants. There were game keepers, fish experts, foresters, gardeners, house attendants, grooms, veterinaries and stable attendants, etc.

Mr. Carnegie was in advance of the times in regard to prohibition He was a total abstainer himself and urged abstinence upon his employees. He gave a 10 per cent. bonus to all the employees at Skibo and in his house in New York every year they reported that they had not drunk any intoxicating liquors. He said once that the first, and most seductive peril and destroyer of most young men, was drinking alcoholic liquors. His views of life were broad and diversified.

After devoting himself to the liberal art of living Mr. Carnegie became a great reader and expressed himself freely on all manner of subjects, economic, social, political, philosophic and literary. He formed many friendships with prominent men in all parts of the world and often in his Fifth avenue house was the host at gatherings of unusual distinction.

He was the author of several books, more than one of which provoked international discussion. His two earliest works were "Round the World," published in 1879, an account of a trip across the Pacific to Japan and back by way of Suez and Europe, and "An American Four-in-Hand in Britain," in 1882. These were followed in 1886 by his best known book, "Triumphant Democracy," which reached in two years a circulation of 40,000 copies, and "The Gospel of Wealth." In 1902 appeared the "Empire of Business," which has been translated into eight languages, and two years later a "Life of James Watt."

Disbursing His Fortune

The clubs, orders, etc., of which Mr. Carnegie was a member include the Lotos, St. Andrews, Riding and Driving, Indian Harbor Yacht, Engineers and the Authors. He was a Commander of the Legion of Honor of France, Grand Cross of Orange and the Grand Cross of the Denebrog. Fifty-three cities of Great Britain and Ireland conferred citizenship upon him.

Mr. Carnegie's benefactions exceeded the sum of $400,000,000. People the world over benefited by the little Scotchman's outpouring of wealth. From New Zealand to California, from Canada to the Orkney Islands, the words "Carnegie Library" have a neighborhood meaning. To the making of many libraries, apparently, there seemed to him no limit save the limits of the globe. He made it the business of his later years to give the opportunity for reading to any community anywhere which manifested a desire for it vital enough to conserve it once it had been given.

He declared that, remembering his own boyhood, when access to a

library of a few hundred volumes was a lighted door of exit from his insufficient early education, he made a resolve to keep that door open to others if ever he had the means. For libraries in this country alone Mr. Carnegie expended upward of $30,000,000. The largest single gifts were to New York, Philadelphia and St. Louis.

Library gifts mounting into the millions were also made to Canada, Great Britain, Scotland, Ireland, Porto Rico, New Zealand, Tasmania and the West Indies. Over $6,000,000 are represented by the libraries Mr. Carnegie gave to England.

Mr. Carnegie had founded 1500 libraries or more and his secretary was considering fresh applications almost every day up to the time of his death. It was Mr. Carnegie's idea that these institutions were in no sense a charity, but a benefit assured by popular support.

"Free libraries maintained by the people," he once declared, "are cradles of democracy, and their spread can never fail to extend and strengthen the democratic idea, the equality of the citizen, the royalty of the man. They are emphatically fruits of the true American ideal."

Much criticism attended the acceptance of these gifts both here and abroad. This was offered on the ground of the great cost of maintaining the libraries, a condition always imposed upon the beneficiaries, and on the ground of the reluctance confessed by some of the captious to conspire with Mr. Carnegie in what was viewed as a vast scheme of self-perpetuation and self-advertisement.

The largest of Mr. Carnegie's single gifts was that creating the Carnegie Institute of Pittsburgh. This gift to the city of his prosperity called for a total outlay for building, equipment and endowment of from $16,000,000 to $25,000,000. The technical schools form a group of buildings near the main structure. It was the technical schools that Mr. Carnegie regarded as of first importance in the foundation. They provide training in the handicrafts, technical methods, and a general industrial education which calls for technical learning or skill. The Institute was formally turned over to the city of Pittsburgh on Apr. 11, 1907, with impressive ceremonies. In his address of presentation Mr. Carnegie said:

"Dollars are only dross until spiritualized, a means to an end, and miserable is the man, mean and squalid his life, who knows no better than to deaden his soul by mere possession, counting over the hoard which holds him down by using his faculties in old age in augmenting the useless stuff which ministers not to any taste worthy of man."

On this occasion occurred a memorable exchange of letters between Mr. Carnegie and Mr. Rockefeller, his great rival in the disbursement of millions. Mr. Rockefeller wrote:

"Please accept my hearty congratulations on your great and good speech at the dedication of Carnegie Institute in your old home city of

Pittsburgh. It has the right ring. I am with you. You have my best wishes for the success of all your grand efforts to help your fellowmen. I hope and trust that our prosperous men the world over will be stimulated to emulate your noble example. I believe that untold good will result therefrom."

Mr. Carnegie answered:

"Many thanks, fellow worker, in the task of distributing surplus wealth for the good of others. I clasp your hand. Your congratulations are highly valued."

Carnegie Foundations Worldwide

Second in size of Mr. Carnegie's benefactions is the Carnegie Foundation for the advancement of teaching and the retirement on suitable pensions of professors of the colleges in this country, Canada and Newfoundland. The amount first set aside for this purpose was $10,000,000 and sectarian institutions were debarred. Later $5,000,000 was added to the original fund.

In January, 1902, Mr. Carnegie called into existence the Carnegie Institution of Washington, a project for the encouragement of higher education and advanced scientific research.

The Carnegie Hero Fund Commission came into existence in 1904 with a fund of $5,000,000 It was designed to make provision for those dependent on persons who had lost their lives or sustained injury in the performance of acts of heroism. Medals were struck bearing Mr. Carnegie's profile in relief which are awarded in recognition of courage of the heroic order. Canada, the United States, and Newfoundland share in the awards of the commission, which has thus far disbursed over $360,000 in benefits and awarded 379 medals.

Mr. Carnegie's partiality for the members of the engineering professions led him several years ago to finance at a cost of $1,500,000 the erection of the United Engineering Building at 25–33 West Thirty-ninth St., New York, which became the headquarters of five separate organizations of engineers.

The number of his gifts to the colleges of the country was legion. Many went to the smaller colleges. To Amherst he gave $100,000 for a new dormitory and twenty-five Carnegie scholarships; the University of Southern California received $25,000; Lake Forest College, $40,000; Western Reserve University, $125,000; Iowa College, $10,000 and a new library; Bates College, $50,000; Brown University, $150,000 toward the John Hay library; Stevens Institute, $125,000; Lafayette College, $50,000; Pennsylvania State College, $150,000; Oberlin, $125,000; Tufts, $100,000; Syracuse, $150,000; Bowdoin, $50,000 for a professorship in history and political science; and Hamilton College, $200,000 toward

the Elihu Root Trust Fund. To Princeton he presented an artificial lake and other gifts.

Education for the Poor a Hobby

Mr. Carnegie was convinced that the problem of the South would be solved by Booker Washington's policy of education. In giving $600,000 for the endowment of Tuskegee Institute he declared that he considered the negro educator "one of the greatest living men, the modern Moses." History, he added, "is to tell of two Washingtons, one white and the other black, both fathers of their people." To Berea College, he gave $200,000 toward the maintenance of a separate school for negro students. To Cooper Union Mr. Carnegie gave $300,000 for the establishment of a mechanical arts day school.

Other dispensations designed to advance the cause of education were $250,000 to the Mechanics' and Tradesmen's Institute of New York, $200,000 for Curtis scholarships to the University of Paris, and $10,000,000 for the Carnegie Educational Fund in Scotland.

In March, 1910, he announced the creation of a fund of $5,000,000 in the interest of teachers and the betterment of educational facilities in ten leading cities of the United States. Of that amount $500,000, yielding an annual income of $12,000, was apportioned to the teachers of New York city.

To the largess publicly given must be added many gifts to hospitals, asylums, friends and old employees. Among the most graceful of his personal gifts was that of the late Lord Acton's library to the Hon. John Morley.

It is noteworthy that Mr. Carnegie never gave money for the support of a church. It was a principle with him not to do so, not because he was himself an agnostic and a student of Herbert Spencer, but because he did not believe the highest good would come of this form of giving. In the donating of church organs, however, he was lavish.

"I am a devoted lover of music," he said only a short time ago. "I give organs to churches or help churches to get organs because I am willing to be responsible for everything the organs say, but I could not be responsible for all that is said from the pulpit."

Spelling reform was another hobby ridden very hard by Mr. Carnegie. Of late years his ardor on the subject perhaps has cooled a bit. His attempts to finance a new orthography into general use had obliged him to submit to a good deal of raillery from those who did not feel the force of the reform argument.

Discrimination in His Gifts

Not all the demands upon him met with assent. He knew how to refuse and sometimes did it with emphasis. He was especially unwilling to give aid where those benefited failed to coöperate with him.

"I have often been asked," he said, "why I do not do so and so for a city or institution. It is usually because I want those places or institutions to help themselves. I love to help those who help themselves."

Worked for European Harmony

The outbreak of the world war came as a great and staggering blow to Mr. Carnegie. For years he had labored hard to establish a peace court, and he gave $10,000,000 to build and maintain a palace at The Hague, where he had fondly hoped all international controversy would be settled without bloodshed. That the nations of the world should ignore his palace at The Hague when the greatest of all peace conference was in progress was another cruel blow.

So earnestly did he strive to reëstablish harmony and understanding between the warring nations that the Germans capitalized his name and incorporated it in their peace propaganda in 1915. Mr. Carnegie denied that he had ever offered money to finance propaganda to bring about a German sought peace.

The efforts of Mr. Carnegie to bring about a world harmony, disarmament and peace by arbitration aroused considerable criticism, but he continued his efforts, established his commissions and redoubled the vigor of his efforts. He did not profess to know how war might be ended, but declared that there must be some way.

When war came to America, however, notwithstanding his great love of peace Mr. Carnegie contributed liberally toward the Liberty Loans. As thoroughly as he had opposed war with Mexico he approved America's declaration of war against the Germans, then threatening to overwhelm Europe with final ruin. He gave with conspicuous generosity toward war welfare work. In the great Red Cross war chest campaign in early 1918 Mr. Carnegie made a single gift of $1,000,000.

The reëstablishment of peace with victory crowning the arms of America made Mr. Carnegie especially happy. He spoke several times in support of the founding of a world league for the preservation of peace, but, when the stress of war had died away Andrew Carnegie, more than fourscore and three, withdrew from the public eye and spent his remaining days absorbed in the more intimate interests of his family, neighbors and friends.

Resolutions of Engineering Societies

At the joint meeting of August 12, the following men were appointed a committee to draft suitable resolutions to be spread upon the records of the bodies represented and sent to the family of Mr. Carnegie: For the Founder Societies and United Engineeering Society: Charles F. Rand, Alex. C. Humphreys, Calvin W. Rice; for the Engineers' Club, Lansing C.

Holden, Robert A Franks, E. Gybbon Spilsbury. Mr. Rand was designated as chairman. The resolutions are as follows:

Andrew Carnegie's death Aug. 11, 1919, at Lenox, Massachusetts, brought to its close a career which greatly advanced all the engineering arts and sciences. By the introduction into the United States of the Bessemer process for the production of steel and by the establishment and development of steel plants, which became the greatest in the world, he made available for engineers the most useful modern material for engineering construction. In the successful conduct of many industrial enterprises, he amassed great wealth, the possession of which he came to regard with deep seriousness as a public trusteeship. He devoted himself to the distribution of large portions of his fortune to projects for the benefit of mankind. He distributed his wealth not only in many directions, but also with the exercise of great wisdom based on careful investigation. His munificence provided large funds for the building of a home for the great national engineering societies and many associate societies. He was an honorary member of the American Institute of Mining and Metallurgical Engineers and American Society of Mechanical Engineers. He was personally known and loved by many engineers. In view of these facts, be it

Resolved, That the American Societies of Civil, Mining and Metallurgical, Mechanical, and Electrical Engineers, the United Engineering Society and the Engineers' Club, herein express to the family of Mr. Carnegie and record their sincere appreciation of the great contributions of Andrew Carnegie to the advancement of engineering, and of his friendly assistance in making possible beautiful homes for the Engineering Societies and the Engineers' Club, thus fostering the spirit of unity in the profession.

Joseph Esrey Johnson, Jr.

JOSEPH ESREY JOHNSON, JR., had already achieved rare distinction as an able metallurgist, clear thinker, brilliant author, and wise consulting engineer to bankers and operators; he had achieved the essentials of a great career when death cut short his activities on Apr. 4, 1919, in the forty-ninth year of his age. He belonged to a family of iron blast-furnace men and mine managers. His father, Major J. E. Johnson, after serving with distinction in the Army of the North, identified himself with the iron mines and blast furnace at Longdale, Va., with which he was connected during almost his entire business life. It is significant that, during those troublous times of reconstruction in the South, he enjoyed always the affection and respect of his neighbors and associates alike. From him his son obtained not only a thorough training in blast-furnace practice and iron mining, but great personal courage and force of character. His mother, who survives both her husband and sons, possesses rare intellectuality, humor and a tenderness and human sympathy which those who were privileged to know well J. E. Johnson, Jr., recognized beneath his exterior of aggressiveness and will power.

In a recently published analysis of the qualities which characterize the world's great men, three traits stand out preëminently; viz., independence, courage, and intellectuality. Each of these characteristics J. E. Johnson, Jr., possessed in extraordinary degree. His early education was obtained under a private tutor at Longdale, and this circumstance probably enhanced a natural tendency to independence of character and individuality which, however, under the control of his unusual power of straight thinking and clear analysis never allowed him to go far astray. It has been said of him that if once he set his mind to a problem, however knotty, he scored a bull's-eye before the subject was dropped. His manner of attacking a subject involved first mastering the details of the available knowledge; then, putting this acquired information to one side, with an untrammeled mind blazing out a path of his own, so clear, so well-defined, and expressed in such simple terms that he illuminated the whole subject without distorting the original data or discussion. His two books, "Blast-furnace Construction in America" and "The Principles, Operation, and Products of the Blast Furnace," on the iron blast furnace together form the most comprehensive, the most enlightening, and the most useful treatise on the subject ever produced in any language. But his ability did not end with the mastery and elucidation of principles enunciated by his predecessors; his own

discoveries and inventions in the metallurgy of iron have established his reputation in every civilized country where iron and steel are made. The three most important of these doubtless are his thermal theory of the iron blast furnace, which ranks with Sir Lowthian Bell's chemical

JOSEPH ESREY JOHNSON, JR.

practice of the iron blast furnace as the two greatest scientific advances in our knowledge of the art; his original classification of iron and steel as an unbroken series of alloys of iron and carbon; and his discovery that oxide, instead of being the unmixed evil which it was accredited in the minds of most metallurgists, actually benefited the strength of cast iron.

There are many stories told of Esrey Johnson's personal courage, but probably the most spectacular one is the prevention of attempted lynching in Virginia in 1909. A young white girl had been murdered by a colored boy under those horrifying circumstances with accounts of which we are all too familiar. Johnson was not only manager of the mines and blast furnace of the locality, but was one of the officials of the county. He organized the search whereby the negro was captured, brought to Johnson's office and a confession of the crime wrung from him. The usual crowd collected quickly, thirsting for vengeance in the form of hanging or burning. Practically alone, Johnson held this crowd at bay for several hours, sometimes mixing with the crowd and pledging his honor that the colored boy should be tried and executed according to law, sometimes pleading for the good name of the community, and when necessary, "showing his teeth" to those who can understand no other argument for the curbing of their personal inclinations or intentions. He has told me that he expressed himself in substantially these words: "I am going to see this nigger hung in accordance with law and order. If you fellows try to prevent me, you may get me, but I will get some of you first." I am sure that at this point in the argument he smiled, for he had a very expressive smile which endeared his friends to him, and which he could make very convincing when he desired. Finally, the morning train arrived on the track just outside Johnson's office. He realized that this was the moment when a coup would be attempted, but he had already laid his plans with the sheriff over the long-distance telephone. The train stopped exactly where he had directed. He rushed the trembling culprit by the crowd of angry, heavily armed men into an empty railroad car occupied only by the sheriff. The door was barricaded behind the three men and train pulled out before a rush could be accomplished. Another crowd had to be over-awed at the entrance of the county courthouse, where the judge had called a special term of the court. To him Johnson delivered his prisoner and made good on his promise that justice should be done, which it ultimately was. The whole story of this interesting episode is told in the issue of the *American Law Review* for November–December, 1911, under the title "Lynching Unnecessary."

Johnson led a white life. There was no weak spot in his professional or personal armor of honor and sincerity. His home life was an unusually happy and congenial one. He married Miss Margaret C. Hilles of Wilmington, Del., and she was not only his friend and comrade, but an intellectual stimulus and helpmate in a very busy life. She, with one boy, J. E. Johnson, 4th, survives him.

I cannot give a better summary of Johnson's professional attainments than by quoting from a letter from his friend, James Gayley. These two master minds found mutual sympathy and understanding. Johnson had an unbounded admiration for Gayley, whom he describes in the

dedication of his books as the "founder of modern American blast-furnace practice." Gayley made of Johnson a pupil and gave him valuable advice and assistance during the early days of his establishment as a consulting engineer in New York City. On the day of Johnson's death, Gayley wrote me as follows:

"American metallurgy of iron has lost its shining light and the world is poorer thereby. . . . Esrey Johnson is and was the greatest man in the metallurgy of iron and the clearest thinker of any man today. He seems like a younger brother gone and none come to fill his place."

Johnson always identified himself with the public work of his profession and community. He was for many years a member of the principal engineering societies, and especially of the American Society of Mechanical Engineers, The Mining and Metallurgical Society of America and the American Institute of Mining and Metallurgical Engineers. His pen has enriched the literature of these and other societies on both sides of the Atlantic. His interests naturally brought him more closely into affiliation with the A. I. M. E. for whom he served as a member of the Board of Directors and chairman and member of many prominent committees, including the vice-chairmanship of the Iron and Steel Committee. Of recent years he has been active in guiding the policies of this Institute, and at the time of his death was, besides a director, a member of the Executive Committee and chairman of the Committee on Admissions.

BRADLEY STOUGHTON.

SKETCH OF LIFE *

Joseph E. Johnson, Jr., was killed accidentally at Hartsdale, N. Y., on the morning of Apr. 4. On the way from his home to the railway station he paused for an automobile to pass. The automobile skidded and struck him, knocking him down and fracturing his skull. He died an hour later without regaining consciousness. The news of this sad accident came as a shock to his host of friends in the profession and outside of it.

J. E. Johnson, Jr., was born at Longdale, Va., in 1870, the son of Major J. E. Johnson, who was the manager of the Longdale Iron Co. He was graduated from Haverford College in 1888. He received the degree of mechanical engineer from the same college in 1891, and in 1892 obtained a degree from Cornell University. In the meanwhile he had begun practical work, first as a draftsman for the Baldwin Locomotive Works. Having finished his studies at Cornell, he was engaged from 1892 to 1894 with the Straight Line Engine Co. and the Cranbery Iron and Coal Co. In 1894–95 he was assistant to the superintendent of the Ames Iron Works, and in 1895–99 was engineer and assistant mana-

* W. R. Ingalls, in *Engineering and Mining Journal*, Apr. 12, 1919.

ger of the Longdale Iron Co. Leaving that position, Mr. Johnson worked with the Carnegie Steel Co. in 1899–1901, being identified with its blast-furnace department. In 1901 he returned to his old position with the Longdale Iron Co. with which he remained until 1906. From 1906 to 1909 he was general manager of the Princess Furnace Co., operating blast furnaces, ore mines and a railroad. During 1909 and 1910 he was general superintendent for the Republic Iron and Steel Co. at Thomas, Ala., operating three blast furnaces and 900 coke ovens.

From 1910 to 1913 Mr. Johnson was manager of the Ashland plant of the Lake Superior Iron and Chemical Co. at Ashland, Wis. In 1913 he opened an office as consulting engineer in New York, and had made this his headquarters since that time. This is a bare outline of his extensive and active career in work among iron mines and furnaces and coking plants. He gained comprehensive experience as a mechanical engineer, as mining engineer and as metallurgical engineer and was splendidly fitted for consulting practice when he entered it.

Previous to his entry into general consulting practice, Mr. Johnson had done much technical work of a major character, which had won for him high professional recognition. His remarkable versatility was exhibited in the variety and wide range of his interests. The introduction of a masterly method of mining, the designing of some ingenious mining machinery, the arrangement of a clever system of tramway transportation, the improvement in blast-furnace practice, the study of the quality of pig iron and the means for raising the grades, were all subjects that engaged his attention with distinctly useful results. He was a generous contributor to the technical press and described his work and offered his ideas in papers in the *Engineering and Mining Journal*, *Chemical and Metallurgical Engineering*, the *American Machinist*, *Iron Age*, and the *Transactions* of the American Institute of Mining Engineers. Besides these contributions, which were invariably noteworthy, he was the author of treatises on "Blast-furnace Construction in America" (1917), followed by "The Principles, Operation and Products of the Blast Furnace" (1918). In his preface to the former work he said: "I happen to have made several investigations which served to explain some of the phenomena of the blast furnace not previously understood, and it seemed proper that the results should be recorded. . . . In many cases my opinions are given for what they may be worth, because it seems to me that those books which simply give a number of opinions without any indication of the author's preference are of little real use." This expression was thoroughly characteristic of the man.

The best efforts of Johnson's leisure for more than three years were spent upon this book, which dealt chiefly with the mechanical side of the blast furnace. It was written in that exact and clear language that was characteristic of its author, which made it easy for everyone to read

and put it within the grasp of any intelligent furnace-man. The succeeding treatise exhibited the same admirable qualities.

In a review of "Blast-furnace Construction" by Bradley Stoughton, than whom no one is more competent to pronounce an opinion, Johnson's professional status was excellently summarized in the following paragraph:

"Probably no living person is qualified for so many different reasons to write a book on this subject as is J. E. Johnson, Jr. His father was an important American ironmaster, and this author may be said to have been born and bred around a blast furnace. For more than 20 years he has held a prominent position as operator, inventor, author, and interpreter. Two of his achievements in the latter capacity would alone be sufficient to place him in the first rank of modern scientific interpreters; namely, his theory of the constitution of cast iron, which has since been expanded and elaborated by himself and others, and his theory of the critical temperature of the blast furnace, which is now generally accepted as the explanation of the saving effected by drying blast, which was once considered (especially by German theorists) as super-theoretical and, therefore, impossible."

In his work as consulting engineer Mr. Johnson found a wide field of activity. One of his recent important works was a professional mission to China. Just previous to his untimely death he had been devoting himself to a new study in the metallurgy of iron, in which he was intensely interested and as to the successful outcome and practical value of which he had great hopes.

Personally, Mr. Johnson was a most lovable character. He had a natural gift for making friends, all of whom enjoyed him immensely. He was a broadminded man, whose interests were far wider than those of his profession. He was a good citizen. He was a wonderfully clear thinker, and his views were invariably characterized by a striking originality. Added to this was his art of frank, clear, and illuminating expression. A sleepy meeting always woke up promptly when Johnson rose to make any remarks. There is nobody to fill the chair that he has left vacant.

Mr. Johnson was a member of the Mining and Metallurgical Society of America and a member of the board of directors of the American Institute of Mining and Metallurgical Engineers, in which he was marked for high honor. He resided in a beautiful old-fashioned home at Hartsdale, where he leaves a widow and a young son.

Resolution of Board of Directors

The resolution drafted by J. W. Richards, J. V. N. Dorr, and Allen H. Rogers on the death of J. E. Johnson, Jr., and adopted by the Board of Directors, is as follows:

"The Board of Directors of the American Institute of Mining and Metallurgical Engineers wishes to place on record its great personal loss in the sudden death of Mr. J. E. Johnson, Jr., our fellow Director of the activities of the Institute, a member of our Executive Committee, Chairman of the Institute's Committee on Admissions, and Vice-chairman of its Iron and Steel Committee.

"A mere recital of these positions of trust and responsibility, which he filled with conscientious efficiency and marked ability, shows how greatly the Institute has lost by his death. No one else of the Institute's officers could have been so illy spared. We, his colleagues, take up his tasks with a sense of our inability to discharge them as well as he, and this with a doubled sense of our loss.

"Others have written and spoken of the blow which his death has been to the furtherance of the metallurgy of iron and steel. His preeminence in this line is one of the facts of which the Institute is justly proud. Here his personality, originality, courageousness in advocating new principles, clearness of writing and forcible presentation of facts, are irreplaceable. His two great treatises on the blast furnace are his literary and scientific memorial.

"We speak but briefly of our personal sense of loss: our incisive but also constructive critic, our helpful organizer, our ever-willing coworker, our cheerful, smiling, optimistic, and most respected friend, our intimate companion, our open-hearted comrade, has 'passed on'—and yet his memory shall be our good cheer and inspiration to further service.

"In the name of the American Institute of Mining and Metallurgical Engineers, we express its sympathy with his bereaved widow and son, and to the outside world we publish this expression of the appreciation of the Institute of his great services to it and of his valuable contributions to the progress of applied science in the industries."

Henry Clay Frick

HENRY CLAY FRICK, a pioneer in modern coke and steel industry and, in more recent years, one of the outstanding financiers of America, died on Dec. 2, 1919, at his home on Fifth Avenue, New York. Although he had been ill since election day, when he had an attack of indigestion which developed into ptomaine poisoning, his death was a surprise to all. He is survived by his widow, Adelaide Howard Childs Frick; a son, Childs Frick, one unmarried daughter, Helen C. Frick, and four grandchildren.

A quiet funeral service was conducted at the Frick home on Wednesday afternoon by the Rev. Dr. Leighton Parks, pastor of St. Bartholomew's Episcopal Church. This was attended by members and very intimate friends of the family, Trustees of the Metropolitan Museum of Art, Officers of the Links Club, and Directors of the United States Steel Corpn.

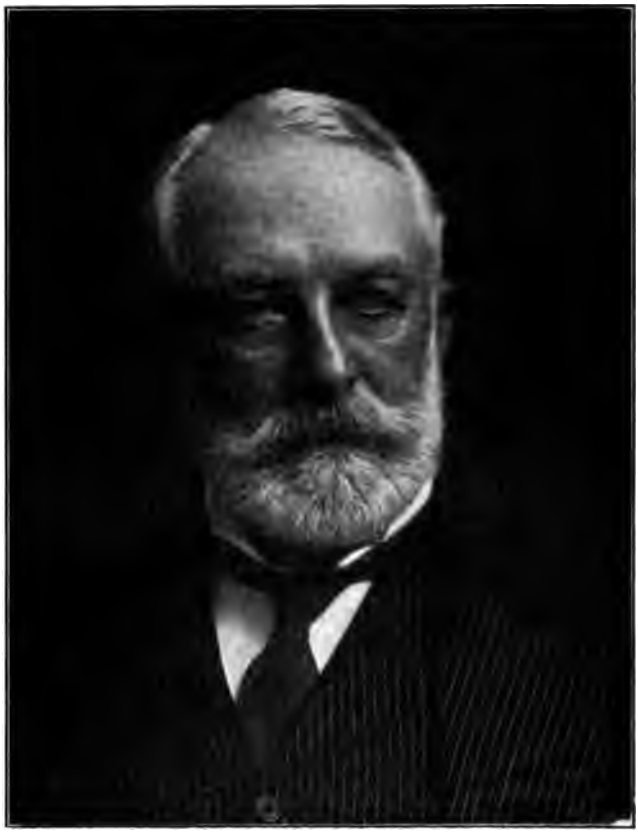
Henry Clay Frick was born Dec. 19, 1849, in West Overland, Westmoreland Co., Pa. His father, John W. Frick, was a farmer of Swiss ancestry, and his mother, Elizabeth Overholt Frick, a member of an old Mennonite family. Until he was sixteen, he spent his time at school, on his father's farm, and in his grandfather's distillery, where he kept books. He attended Otterbein University, in Ohio, for a year.

Frick, learning that coke was an essential of the already rapidly developing steel business, early invested every cent he could get in coking-coal lands. With the help of an associate of his grandfather, he formed the corporation of Frick & Co., coke dealers, and acquired fifty-one ovens in the Connellsville region and 300 acres of soft-coal lands. During the panic of '73, with the help of Judge Thomas Mellon, a Pittsburgh banker, he bought out his partners and while coke was selling at 90 cents a ton enlarged his purchases of suitable lands. Later, the price of coke increased until it was selling at $5 a ton. Before he was 30 years old, Frick was rated a millionaire.

ASSOCIATION WITH CARNEGIE

In 1878, he sold an interest in the business to E. M. Ferguson of New York, and later a share to Mr. Ferguson's brother. Four years later the business was reorganized into the H. C. Frick Coke Co. with vast coal lands, great banks of ovens, and a capitalization of $2,000,000. Then began his association with Andrew Carnegie. He exchanged a part of his coke interest for shares in Carnegie Brothers of Pittsburgh,

the steel company and the coke company having become very necessary to each other. The capital stock was increased $1,000,000, and within three years Carnegie had gained a majority control by buying out the Fergusons. Two years later Mr. Carnegie picked Frick as chairman of Carnegie Brothers & Co., Limited, into which his firm had been reorganized.



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HENRY CLAY FRICK

Several years later Frick, with Henry Phipps, Judge Moore, and others, got from Carnegie an option on his holdings at a price in excess of $157,000,000; for this he paid $1,000,000 in cash. He induced the late J. P. Morgan and others to visit Pittsburgh and tried to make them see the great future of the industry as he saw it. The death of Governor Roswell P. Flower while the deal was under discussion, with a resultant depression in Wall Street, caused the plan to fail. Later, Charles M. Schwab, with Carnegie's aid, put through the deal at a higher price than

Frick had asked. By this deal, Frick found his holdings enhanced to a value of some $61,000,000.

In 1907, Frick, already credited with owning Pennsylvania Railroad stock valued at $8,500,000, executed a coup in the stock of the Reading Railway, which he controlled, whereby he was said to have made from $4,000,000 to $6,000,000. That was the last great deal in which he figured publicly.

He was a student and lover of art, and by the use of patience and thought and large sums of money he formed one of the finest private collections of paintings, statuary, bronzes, porcelains, enamels, furniture, and other objects of art, all of which, under the provisions of his testament, will in due time be permanently turned over to the public use and enjoyment, together with his costly home in New York, adequately endowed.

Samuel Thomas Wellman

SAMUEL T. WELLMAN, Cleveland pioneer steel man, who was often referred to as the "father of the open-hearth process of the United States," died suddenly on July 11, 1919, of heart disease, at Stratton, Me. For more than half a century, he was closely identified with the improvement of equipment used in the making of steel and contributed a number of inventions that virtually revolutionized the industry.

Mr. Wellman was born Feb. 5, 1847, at Wareham, Mass., his father at the time being superintendent of the Nashua Iron Co., Nashua, N. H. After attending public school and a year at Norwich University, he served in the Civil War as a corporal of the first New Hampshire Heavy Artillery. He went to Pittsburgh in 1867 and assisted in starting the first crucible steel melting furnace built in America at the works of Anderson, Cook & Co. It melted a ton of steel with an average of 1000 lb. of nut coal, which cost less than $1, while the melting of a ton of steel in crucibles in the old fashioned coke furnace took three tons of the very best coke, costing from $2 to $3 per ton. This furnace was a great success and in a very few years had driven the coke furnace out of use. From Anderson, Cook & Co., Mr. Wellman went to the Singer, Nimick & Co. works, Pittsburgh, where he built two crucible steel melting furnaces of the same type. After that he spent some time in the office of the Siemens agents in Boston and also at steel works in different parts of the country, starting crucible steel furnaces. He then went to the Bay State Iron Works in South Boston, Mass., where he built the first open-hearth furnace that was a commercial success in the United States. It made steel of a much better quality than had previously been made in this country. The principal use to which it was put in those days was the manufacture of locomotive fire-boxes. From South Boston, he went back to the old works in New Hampshire where his father was still superintendent, and built for it an open-hearth furnace, a plate mill, and a bar mill.

In 1873, Mr. Wellman went to Cleveland to design and build the plant of the Otis Steel Co. He remained with this company for 16 years as engineer and superintendent. During this time he developed the open-hearth charging machine and introduced the electromagnet for handling pig iron and scrap steel. The first charging machine was operated by hydraulic power, and although it was not entirely satisfactory it performed the work fairly well and encouraged its inventor to build another one, which was driven by electric motors. This proved

more satisfactory, and was installed in the plant of the Wellman Steel Co., Thurlow, Pa., a company organized by Mr. Wellman in 1890. A second electric machine was installed in the plant of the Standard Steel Works, Burnham, Pa. The machines at Thurlow and Burnham were of the low type. In 1895, four high-type machines were built for the Homestead works of the Carnegie Steel Co., and from that time the development and

SAMUEL T. WELLMAN

use of electric charging machines has been rapid. Charles M. Schwab once stated that he believed that the open-hearth charging machine effects a saving of 50 cents per ton of steel. Inasmuch as upwards of 300,000,000 tons of open-hearth steel have been manufactured since the conception of the charging machine, Mr. Wellman's invention has been an important factor in the industry. His machine for charging open-hearth furnaces with white-hot steel was his greatest contribution to steelmaking and has been adopted by the steel concerns of the world, earning the inventor international fame. Machinery built under his

patents is being used in Great Britain, France, India, Japan, Russia, Spain, Egypt, and Germany.

In 1896, Mr. Wellman organized the Wellman-Seaver Engineering Co. in which his brother, Charles H. Wellman, and John W. Seaver were associated. Later this company was consolidated with the Webster, Camp & Lane Co., Akron, O., forming the Wellman-Seaver-Morgan Co. This company retains the name at the present time, although Samuel T. Wellman retired from active participation in the year 1900.

He was a member of many engineering and scientific societies, including the American Society of Mechanical Engineers, of which he was president in 1901; American Institute of Mining and Metallurgical Engineers; Iron and Steel Institute of England; American Iron and Steel Institute; and the Cleveland Engineering Society, of which he was an honorary member.

He is survived by two daughters and three sons; of the latter, all living in Cleveland, W. S. Wellman is president of Wellman Products Co., and M. C. and F. S. Wellman are connected with the Wellman Bronze Co.

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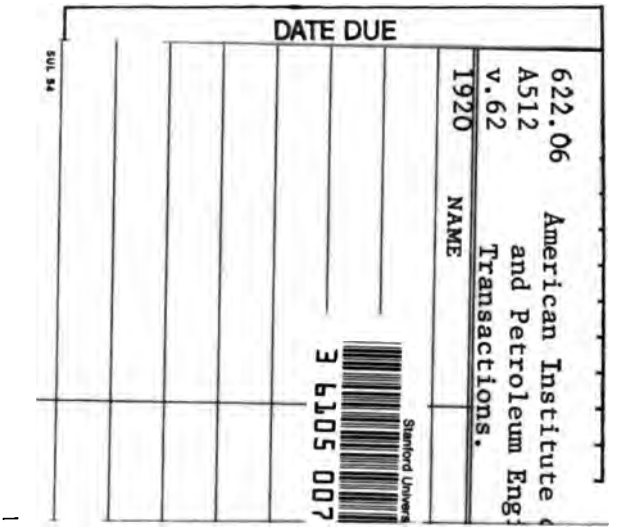

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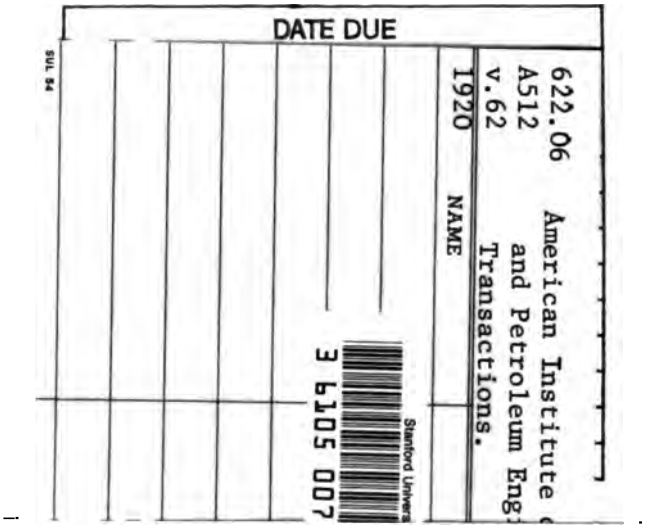
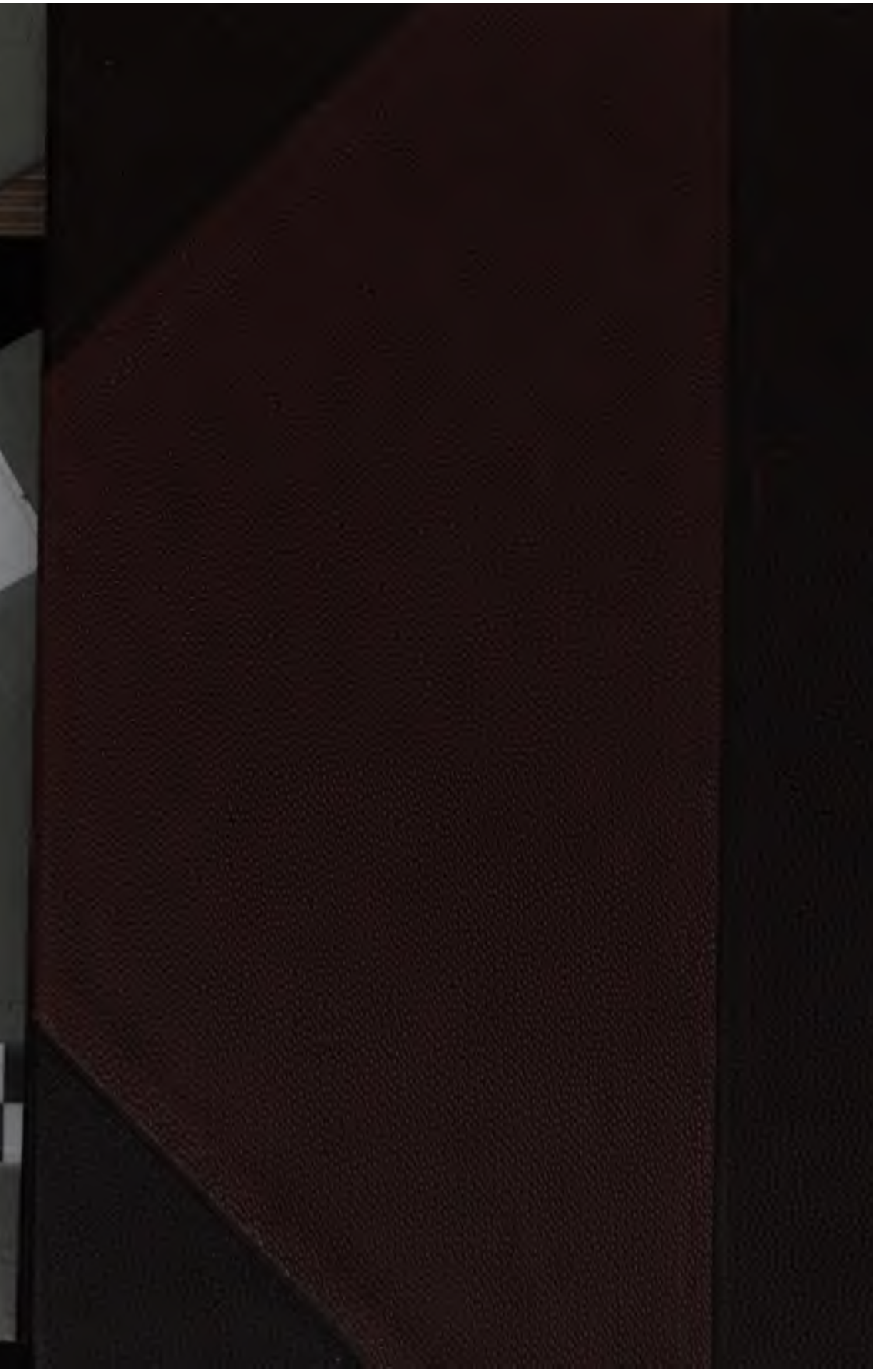
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